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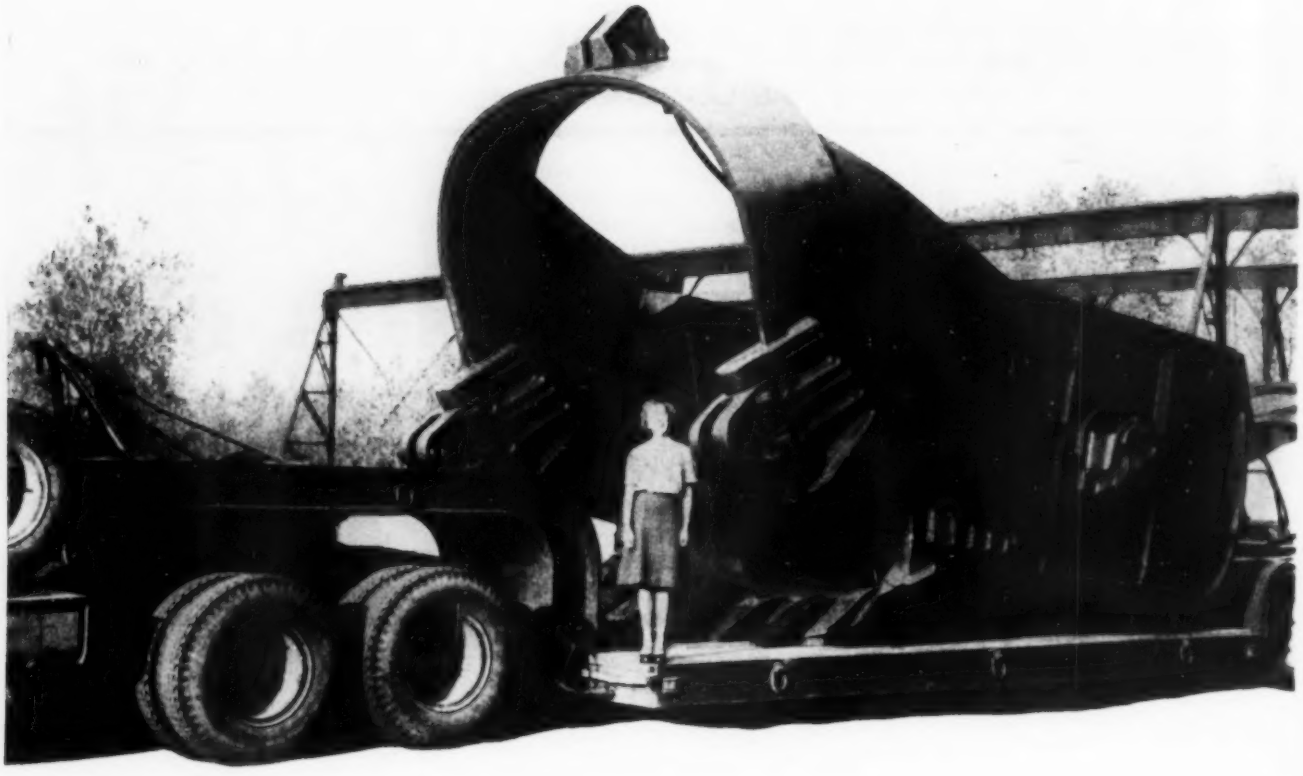

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Subzero Treatments for Carburized Medium-Alloy Parts

CONTROL of the austenite content of the case of carburized and hardened parts is a comparatively simple matter, the principles of which are quite well known. The amount of austenite retained is a function of the carbon and alloy in solution at the instant of quenching and the direction and intensity of the internal stresses induced in the particular part by this operation. Since little or nothing is known about the latter factor, it is generally assumed to be constant for a given part, quenching operation and case depth. Austenite control in a given alloy steel, then, resolves itself into selection and strict control of the maximum carbon content and carbon distribution of the case in carburizing, and of the quenching temperature, speed of quench, and time in the hardening operation. By fulfilling these conditions, the full range of alloy steels of carburizing grades may be satisfactorily casehardened to produce a case with low austenite content and with sufficient excess carbide for excellent wear resistance and a minimum surface hardness of C-58.

The above remarks are directed toward those parts which may be ground after hardening, so that distortion inherent in the double quenching treatment can be corrected. When the distortion must be held to a minimum because the part must operate in the unground condition, as in automotive gears, the choice of heat treatments is considerably limited. The lower alloy steels are used as-quenched from the carburizer, with or without a subsequent temper, and are satisfactory in this condition. At the higher alloy levels, however, the austenite content of the as-quenched gear is often too high for a satisfactory hardness and structure in the single quenched condition. The alternatives are *(a)* cool slowly from the carburizer (followed by a reheat and quench) and accept the carbide network in the hypereutectoid zone, or *(b)* double treat and become resigned to a considerably longer lapping time to correct the additional distortion.

Steels and Tests

In the hope that subzero treatment of single quenched, medium-alloy carburized material would reduce the austenite content and hence improve the hardness and structure, the experimental work described herein was undertaken. Effects of various subzero treatments on the hardness, microstructure and size change were studied on seven common alloy carburizing steels in the single quenched condition: S.A.E. 2512, 4620, 4820, 8620, 4320, 3310 and Krupp. Although retained austenite is not a great problem in gears manufactured from S.A.E. 4620, 8620, or 4320 steels, they were included in this study as a matter of interest, as they are often used after quenching from the pot and tempering. In addition, it may be desired under some con-

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Table I — Chemistry of Steels Studied

Type	Diameter of Bar	C	Cr	Ni	Mo
2512	2½ & 1¼ in.	0.105	0.15	4.95	0.03
4620	2½	0.205	0.24	1.83	0.23
4320	2½ & 1¼	0.21	0.47	1.84	0.23
3310	2½	0.13	1.33	3.65	0.04
3310	1¼	0.10	1.38	3.56	0.03
Krupp	2½ & 1¼	0.13	1.51	3.94	0.03
4820	2½ & 1¼	0.185	0.21	3.43	0.26
4620	1¼	0.20	0.20	1.74	0.23
8620	2½ & 1¼	0.20	0.50	0.55	0.22

ditions to lessen the austenite in the case of a gear made of these steels.

All samples were from production heats of electric furnace and openhearth steels. The analyses of the heats are listed in Table I.

Hardness and microstructure determinations were made on 1-in. rounds, 1¼ in. long, cut from 1¼-in. diameter hot rolled bars. Rings for measuring size changes and carbon penetration were machined from 2½-in. forged bars. The carbon test ring is 2½ in. outside diameter, 1½ in. inside diameter, 1½ in. wide. It is carburized with the charge, mica cooled, and machined in 0.005-in cuts for subsequent chemical analysis.

Samples for hardness and microscopic tests (except for S.A.E.8620) were carburized in a raw natural gas atmosphere at 1700° F., using a 19½-hr. total time cycle, composed of 4½ hr. heating time, 10 hr. carburizing time, and a 5-hr. diffusion period. Upon discharge, the 1-in. diameter by 1¼-in. long samples were quenched with the rest of the charge in oil at approximately 150° F., and the carbon test rings were mica cooled to permit machining. Results of the analyses showing carbon penetration are given in Table II.

After quenching, the bar samples were divided into two lots, one of which was given an immediate temper for 1 hr. at 350° F. The other lot was subdivided further for three cold treatments at −100, −80 and −40° F., respectively (1 hr. each). Since the quenching oil was at about 150° F., all samples could be got into the various cold treatments before they had reached room temperature, thereby approximating a continuous quenching from 1700° F. to the respective subzero temperatures. A 350° F. temper for 1 hr. followed the cold treatment.

The lot which was tempered immediately after the quench was then treated as noted above for the other lot with the exception of one set of samples, which was retained in the quenched and tempered condition. All samples were in duplicate.

After all heat treatments were completed, flats were ground on diametrically opposite sides of each bar, one 0.005 in. below the tangent surface and the opposite one 0.010 in. below. Ten Rockwell hardnesses were taken on both flats, and an average computed. When the hardness results for the two levels were compared, no significant difference was noted on any of the steels, so the two sets of values were averaged and are reported hereinafter as an average value for each treatment.

After these initial hardness determinations, the duplicate samples were given additional cold treatment at the temperatures previously used. (The longer times at −80° F. were omitted, as this temperature could not be controlled with sufficient accuracy for the desired lengths of time.) The additional treatments were done in two steps, namely, 5 hr. (for a total of 6 hr.) followed by hardness determinations, and 18 hr. (a total of 24 hr.) also followed by hardness tests. The results were treated exactly as were the first set of hardness values. Table III summarizes the treatments used, and the hardness values obtained are shown graphically in Fig. 1 and 2. S.A.E.8620 steel (Fig. 1) was not included in the original complement. When it was decided to include this analysis, some of the original test conditions had been proved valueless and were omitted in this series.

Table II — Carbon Penetration on Test Rings

Depth Below Surface	Carbon Content of Steel Noted						
	2512	4620	4820	4320	3310	Krupp	8620
0.005 in.	0.96%	1.00%	1.03%	0.97%	0.92%	0.97%	0.92%
0.010	0.97	1.05	1.03	1.00	0.94	0.99	0.96
0.015	0.94	1.03	1.01	0.99	0.93	0.99	0.93
0.020	0.92	1.02	0.99	0.96	0.92	0.96	0.89
0.025	0.89	1.00	0.93	0.92	0.86	0.92	0.87
0.030	0.85	0.94	0.89	0.90	0.83	0.89	0.82
0.035	0.82	0.90	0.85	0.85	0.81	0.85	0.70
0.040	0.78	0.86	0.81	0.82	0.75	0.81	0.74
0.045	0.71	0.82	0.77	0.78	0.71	0.77	0.68
0.050	0.67	0.78	0.72	0.72	0.68	0.72	0.63

Hardness Changes

Since the results showed major changes in hardness, indicating significant variation in the amounts of retained austenite, a microscopic study was made of the samples and typical results are illustrated in Fig. 3 to 5. All photographs were taken at that point in the case which showed the

maximum amount of retained austenite, and the position of this point relative to the surface is noted in the respective captions.

All oil quenched steels showed a hardness increase after any of the subzero treatments. Samples cooled to −100° F. (the lowest temperature used, represented by Code 1 and Code 4 for each steel) showed the highest hardness, either singly or in average. Cooling to the other temperatures investigated also increased the hardness, but usually to a lesser extent. As expected, the effect of retreatment for longer times was not significant. The effect of a temper between quench and subzero treatment is shown to be very slight — compare Code 1 to Code 4, 2 to 5, and 3 to 6. These remarks apply to all steels, although with higher alloy content and lower initial hardness, the changes are proportionately greater.

For evaluating the response of the different steels to cold treatment, they are best divided into two groups: First is the lower alloy steels shown in Fig. 1 (S.A.E.8620, 4620 and 4320), all of which were well above C-60 after cold treatment; second is the higher alloy group shown in Fig. 2 (S.A.E. 2512, 4820, 3310 and Krupp), of which only occasional samples reached C-60 after cold treatment.

Microstructures are shown in Fig. 3 to 8. Three typical sets "before and after" are shown, each of which shows to what extent a different steel may be changed by the subzero treatment.

Table III — Schedule of Treatments After Quenching*

Sample (Code)	Temper†	First Cold Treatment†	Temper†	Subsequent Cold Treatment
1A	None	−100°	350°	No additional
1B	None	−100°	350°	−100°, 5 hr., total 6 hr.
1C	None	−100°	350°	−100°, 23 hr., total 24 hr.
2A	None	−80°	350°	No additional
3A	None	−40°	350°	No additional
3B	None	−40°	350°	−40°, 5 hr., total 6 hr.
3C	None	−40°	350°	−40°, 23 hr., total 24 hr.
4A	350°	−100°	350°	No additional
4B	350°	−100°	350°	−100°, 5 hr., total 6 hr.
4C	350°	−100°	350°	−100°, 23 hr., total 24 hr.
5A	350°	−80°	350°	No additional
6A	350°	−40°	350°	No additional
6B	350°	−40°	350°	−40°, 5 hr., total 6 hr.
6C	350°	−40°	350°	−40°, 23 hr., total 24 hr.

*All quenched in oil at 150° F. from carburizing cycle (1700° F.)
†Temperings and first cold treatment all for 1 hr.

Size Changes

Size changes were studied on rings of approximately 2.18 in. outside diameter, 1.46 in. inside diameter, and 0.365 in. wide, made from the four higher alloy types. The rings were carburized at 1700° F., using a 6-hr. carburizing time followed by a 3-hr. diffusion period, and then quenched from the carburizing furnace. All rings were given a 350° F. temper for 1 hr. and stability tests were begun at this time.

Samples were ground and lapped after the pot quench and temper, and a set of measurements taken on each ring to the nearest 0.000,05 in. Three rings of each type were studied — a total of 12 sets

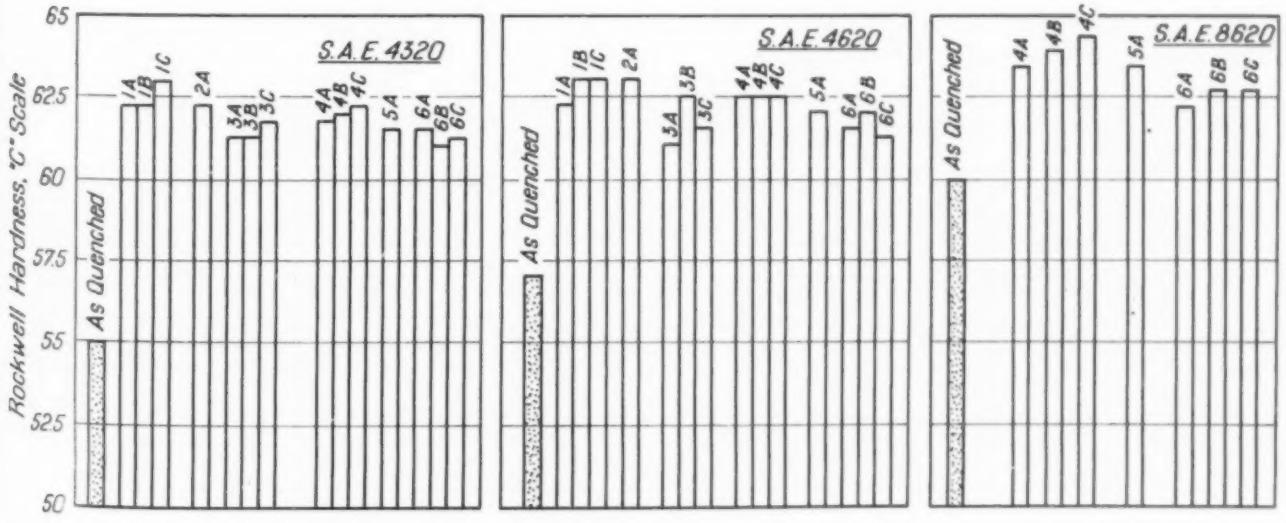

Fig. 1 — Hardness of S.A.E. 4320, 4620 and 8620 After Carburizing, Oil Quenching and Various Subzero Treatments. (See also Fig. 2)

of readings. These measurements were completed slightly more than three months after the carburizing treatment. All samples were then given a subzero treatment (−100° F. for 24 hr.) and returned for final measurements, which were completed one month later. These delays were not intentional, but were due to the fact that the demands of production took precedence over this experimental work.

The growth for all samples was then calculated, and an average value for each steel obtained. The results were fairly consistent. The figures were reduced to inches per inch of outside and inside diameters, and the resulting data are presented in Table IV.

Table IV — Dimensional Changes After Subzero Treatment

Steel	Change, In. Per In.	
	Outside Diameter	Inside Diameter
4820	+0.0052	+0.0045
2512	0.0044	0.0041
3310	0.0074	0.0048
Krupp	0.0062	0.0048

The results are generally in line with the hardness data. That is to say, the steels which showed the largest amounts of austenite after quenching showed the largest change in dimensions after cold treatment — obviously due to transformation of the austenite by subzero treatment. The degree of response is somewhat surprising in view of the long period at room temperature between quench and deep freeze (over three months) since some stabilizing effect might be expected from this holding time.

The data of Table IV indicate a greater divergence between the growth of outside diameter, as compared to growth figures on inside diameter, with increasing austenite content prior to cold treatment. This is believed due to the circumstance that the region containing the austenite has only a very small radial dimension and hence exerts its major influence circumferentially. It is only at the higher initial austenite contents that the growth in the radial direction becomes significant, and causes the inside diameter to grow much less than the outside diameter. Because of the many other factors involved, the data of Table IV should be considered only as confirmation of the hardness changes and not as a guide to the size changes which may take place in production parts.

Conclusions

The following conclusions appear justified from the data obtained:

1. Quite significant increases in hardness can be obtained by subzero treatment of alloy carburizing steels quenched from the

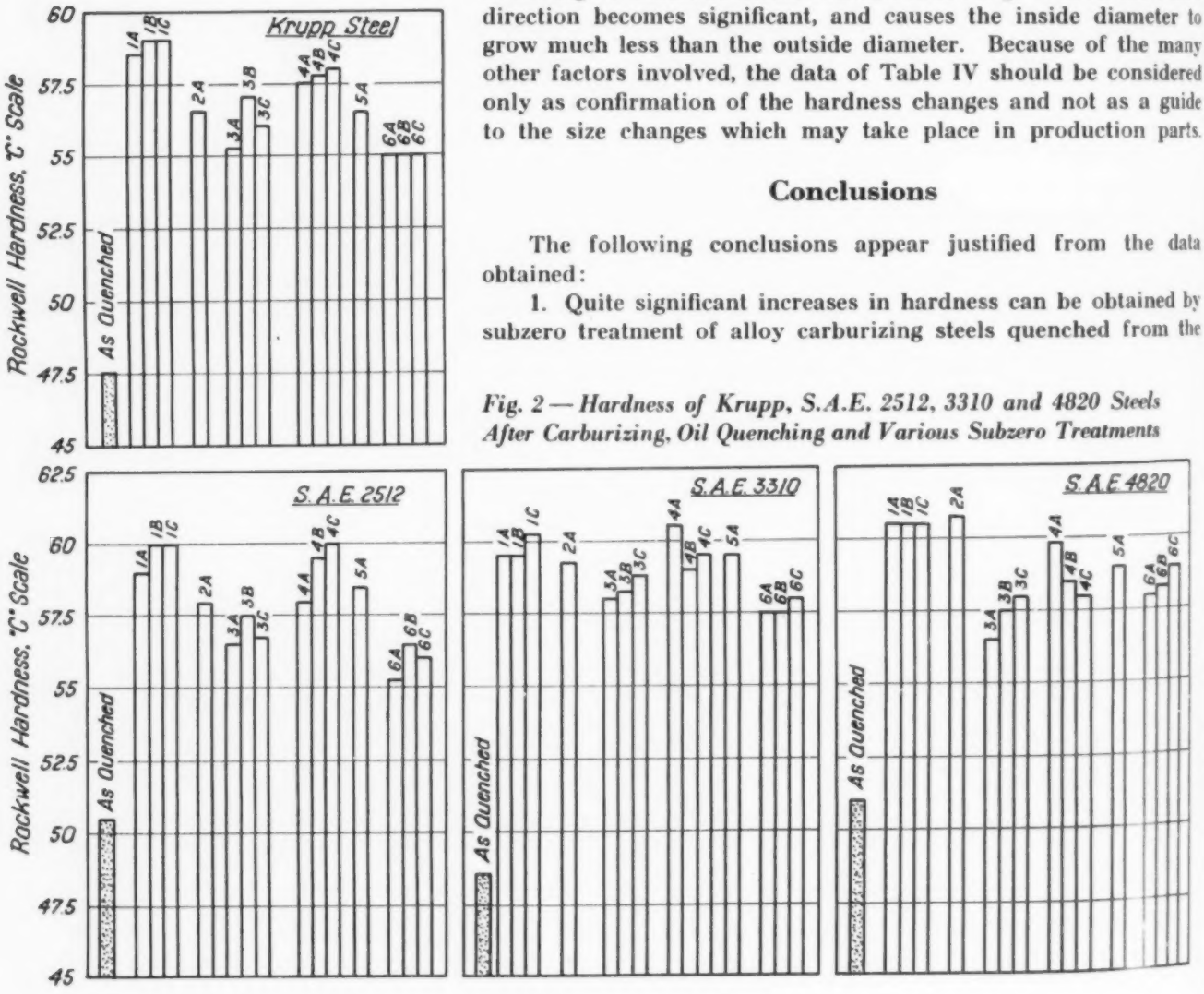

Fig. 2 — Hardness of Krupp, S.A.E. 2512, 3310 and 4820 Steels After Carburizing, Oil Quenching and Various Subzero Treatments

pot. The steels do not appear to be readily susceptible to "austenite stabilization", and may be tempered prior to deep freezing.

2. The choice of the subzero temperature to be used depends upon the steel, the final case hardness desired, and economics. The small increment of hardness obtained by cooling to −100° F. over that obtained at −80 or −40° F. often does not warrant the added expense.

3. No evidence was obtained which would lead to the belief that the product of the subzero treatment was better or different in any way from that obtained by double quenching. Due to the expense, subzero treatment does not seem to offer any advantage in the processing of carburized and ground products, such as roller bearing parts. However, if minimum distortion is of paramount importance in production, deep freezing does seem to offer some hope of producing single quenched parts from medium-alloy steels, having surface hardnesses and microstructures substantially equivalent to those obtained by double treatment, and without the added distortion inherent in the double quench.

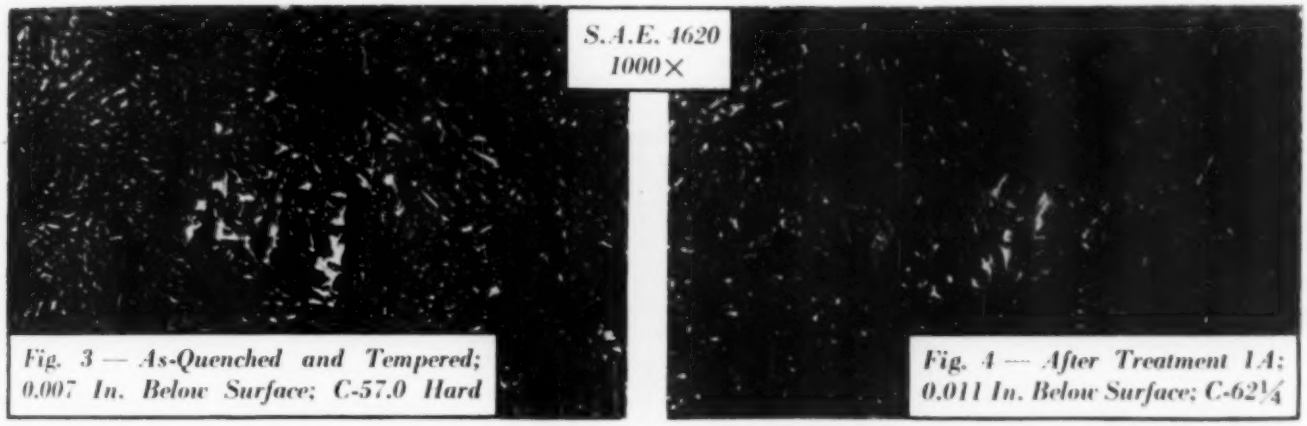

Fig. 3 — As-Quenched and Tempered; 0.007 In. Below Surface; C-57.0 Hard

Fig. 4 — After Treatment 1A; 0.011 In. Below Surface; C-62¼

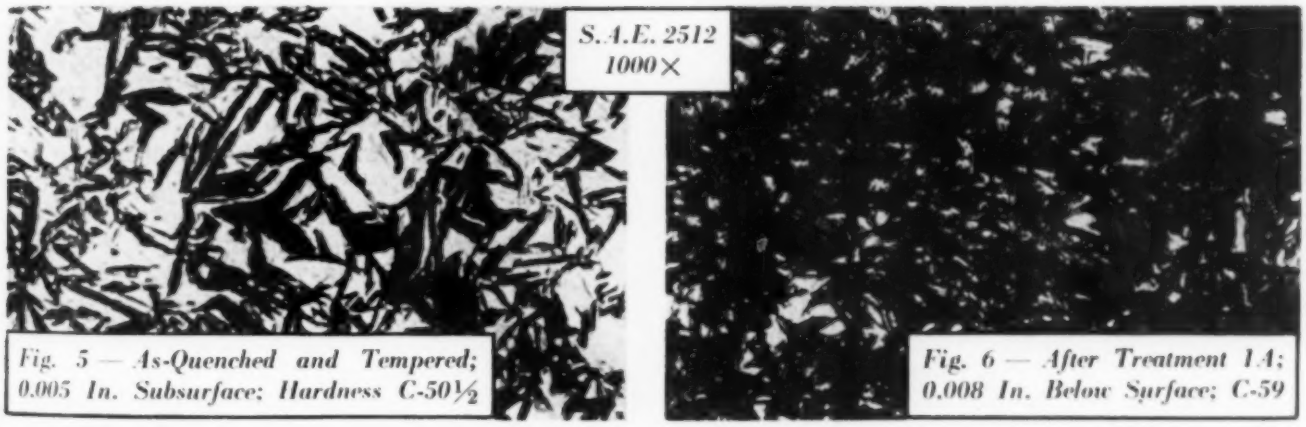

Fig. 5 — As-Quenched and Tempered; 0.005 In. Subsurface; Hardness C-50½

Fig. 6 — After Treatment 1A; 0.008 In. Below Surface; C-59

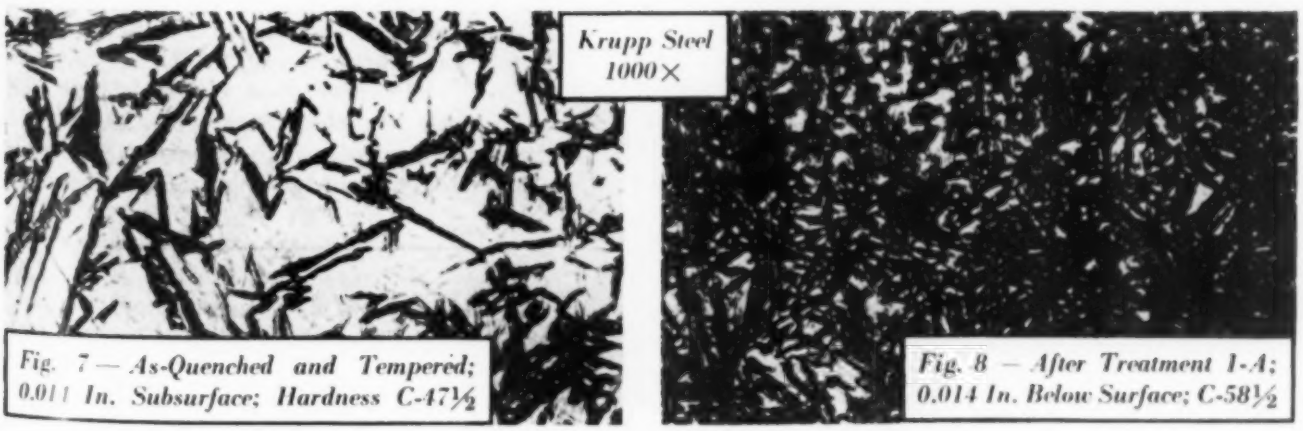

Fig. 7 — As-Quenched and Tempered; 0.011 In. Subsurface; Hardness C-47½

Fig. 8 — After Treatment 1-A; 0.014 In. Below Surface; C-58½

This is the concluding portion of a general article on "Determination of Phase Diagrams" prepared for the 1948 edition of "Metals Handbook". It discusses auxiliary experimental problems and gives a simplified derivation of the phase rule.

Phase Diagrams and the Phase Rule

IN THE FIRST PART of this general article about the determination of phase diagrams was described the various ways whereby the boundaries of the phase fields can be fixed. In this concluding portion certain methods and precautions for such auxiliary operations as temperature measurement, selection and analysis of the representative alloy compositions, and the plotting of the phase diagram will be briefly discussed. Finally, a brief mathematical derivation of the phase rule will be presented, since most of the metallurgical texts treat this matter in a very sketchy way and the treatment in books on thermodynamics is usually entirely too involved. It is thought that if the metallurgist understands the derivation, the phase rule can be applied much more intelligently.

Temperature Measurement

Temperature measurement is usually best done by a good potentiometer which measures the voltage across a calibrated thermocouple in good thermal contact with the specimen. Above the range where thermocouples can be used, a radiation (or optical) pyrometer must be used, making it important to approximate black-body conditions as nearly as possible by sighting down a tube into a relatively large isothermal region in which the specimen is located.

In general, the temperature should be held constant over a sufficiently long period to insure both thermal and chemical equilibrium. However, sometimes it is necessary to take measurements while the specimen is being heated or cooled; temperature gradients should then be minimized by using a specimen of small cross section. Otherwise, inaccuracies in the temperature measurements may be caused by some portions of the specimen transforming at a different time from those adjacent to the couple. It is desirable to record constantly changing temperatures with an accurate, rapid, self-balancing potentiometer.

Chemical composition should be determined by accurate analysis of representative samples. A few precautions will be pointed out:

It is important to use metals of high purity. Very small amounts of impurities may make substantial differences in the diagram. For example, a few hundredths of a per cent of iron in aluminum-manganese alloys change greatly the position of the solvus curve, because the iron goes into solution in the manganese constituent and reduces its solubility. Since it cannot be predicted in advance which elements will cause substantial changes, and which negligible ones, it is essential to use materials of the highest purity obtainable, melted under conditions which will minimize contamination, and the alloys should then be analyzed to reveal any accidental additions.

Alloys should be cast under conditions giving a high rate of progressive solidification, to obtain a sound, fine-grained structure as free as possible from segregation. All ingots or slabs should be scalped to remove any exudation or surface contamination. Whenever the alloy is workable, samples should be rolled, forged, swaged, extruded or otherwise worked to break down the as-cast structure and assist in homogenizing the material.

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Often changes in composition occur during the heating to attain equilibrium. For example, alloys of mercury, cadmium, sodium, zinc, and magnesium may lose these volatile elements during prolonged heating. Some of the specimens actually used in the equilibrium studies should be checked to see whether any such loss has or has not occurred.

If small chill-cast samples are used for the routine analysis, the composition of occasional equilibrium specimens should be checked. It is especially important to analyze specimens giving data inconsistent with those of adjacent specimens.

Obtaining Equilibrium — Equilibrium is established most rapidly in cold-worked specimens from fine-grained ingots. It is often necessary to keep the specimen at the desired temperature for some time, then to quench, cold work, and reheat. Particularly at the lower temperatures, it has even been necessary to repeat this cycle several times to obtain equilibrium. Material which cannot be worked should be chill-cast as small specimens to obtain as fine a structure as possible, and should be held at temperature for prolonged periods. Even under the most favorable conditions, equilibrium cannot be established in cast material at as low a temperature as it can in wrought material.

For some of the points on the diagram, equilibrium should be attained by approach from both higher and lower temperatures. If this is not done, a steady or metastable state may be mistaken for equilibrium.

Selection of Alloy Compositions

Binary Systems — In general the selection of alloy compositions should not be made on the basis of obtaining equal spacing across the diagram. There is no way of determining, before any work is done on the system, exactly what compositions should be selected. Some preliminary work is usually necessary to get a general idea of the diagram. X-ray and microscopic examination of a few cooling curve slugs will ordinarily allow a satisfactory selection of alloy compositions for a more detailed study of the system, or of those portions which are of immediate interest.

Ternary Systems — Mapping a ternary or more complex system is usually a long and arduous task. Every effort should be made, therefore, to reduce the work by proper selection of compositions. The method proposed by Reinders and Guertler in 1915 for the preliminary study is very useful; it has been described in English and extended to more complex systems by Beck (*Journal of Applied Physics*, V. 16, 1945, p. 808).

In ternary systems all the points representing single phases in the binary boundary systems are connected by all possible straight lines. Compositions corresponding to the intersections of some of

Photo courtesy Saginaw Malleable Iron Division, General Motors Corp.

these lines are then used for preliminary examination. Guertler originally recommended microscopic identification of the phases in these alloys, but powder patterns give more positive information. The phases found in these samples show which of the lines represent binary systems and which ones have no physical significance. This permits the large triangle representing the entire system to be divided into smaller triangles representing partial ternary systems.

The above procedure requires some modification whenever the powder patterns reveal the existence of a ternary phase or phases which do not exist in any of the binary boundary systems. Then it is necessary first to determine the composition of this phase and second to connect or relate its composition with compositions of the various binary phases.

The next step in the study of the ternary system is to determine the diagram for those binary systems which have been established by Guertler's method. Then the diagram for each of the partial ternary systems can be drawn after a preliminary survey, as in the binary systems.

Drawing of the Phase Diagram

Weighting of Results — As remarked at the outset, most of the features of an equilibrium diagram should be determined by more than one method. Results obtained should check one another, giving added confidence in the accuracy of the diagram. However, when significant discrepancies occur, the values obtained by the different methods should be weighted in accordance with the expected accuracy. In some instances one set of values may be disregarded entirely — for example, a liquidus curve determined by the settling method and by cooling curves should be drawn through the points determined by the settling method if there is a significant difference between the two sets of values, and if the cooling-curve data are at lower temperatures. (Even after correcting for undercooling, cooling curves may give low results.)

Three-Dimensional Models

In determining binary diagrams, rectangular coordinates are of course used for plotting the data and selecting compositions for additional tests. However, in determining ternary equilibrium diagrams, satisfactory representation of the data is not so simple. Perspective drawings of three-dimensional models or isothermal sections of such models can be used, but unless the investigator is unusually good at visualizing the three-dimensional diagram, it is more satisfactory to construct a space model as the work proceeds.

A simple and fairly satisfactory procedure is to use a base of soft wood on which is ruled the Gibbs's triangle. Stiff pointed wires can be driven vertically into this base to represent the compositions of the samples investigated. Marks of various colors of paint (representing the solidus, the liquidus, and other reactions encountered in the system) are made on these wires to indicate temperatures. Other types of three-dimensional models have been described in the literature.

Thermodynamic Relations

In spite of all the care which may have been used in the experimentation, the results should be scrutinized carefully for errors or inconsistencies. Thermodynamic relations help greatly in such appraisals. The most fundamental of the thermodynamic relations, known as the phase rule, defines the *number* of phases which can be in equilibrium. If the number observed is inconsistent with the phase rule, either equilibrium was not attained in the experimental work or there have been too few experimental points to reveal all of the phase boundaries.

The phase rule is a simple equation which applies to all systems in equilibrium and gives the relation between the number of phases, the number of degrees of freedom, and the number of components. It is desirable to define these terms, as briefly as may be.

The term "phase" was defined in the first paragraph of the article in the April issue. (The air above the crucible, regardless of whether or not it contains some metal atoms, is *not* a phase of the alloy.)

The "number of degrees of freedom" equals the number of intensive variables (properties that are independent of the mass of the sample) which may be altered arbitrarily within certain limits without causing the disappearance of any phase or the formation of a new phase. The intensive variables commonly used in applying the phase rule are pressure, temperature, and the ratios of the masses of the components existing in the different phases.

The "number of components" is equal to the smallest number of independently variable constituents by means of which the composition of each phase participating in the state of equilibrium can be expressed.

The phase rule can be developed very simply if we accept the conditions for equilibrium under constant temperature and pressure (which can be proved thermodynamically by the use of the func-

tion known as free energy or thermodynamic potential), namely, that the temperature, pressure, and the chemical potential of any component is the same in all of the phases. Let F equal the degrees of freedom, C the number of components, P the number of phases, μ the chemical potential, and m the mass of a component. Different phases are distinguished in the equations by accents and the different components by subscripts. Using these symbols, the intensive variables are:

Pressure
Temperature

$$\frac{m_2'}{m_1'}, \frac{m_3'}{m_1'}, \frac{m_4'}{m_1'} \cdots\cdots \frac{m_C'}{m_1'}$$

$$\frac{m_2''}{m_1''}, \frac{m_3''}{m_1''}, \frac{m_4''}{m_1''} \cdots\cdots \frac{m_C''}{m_1''}$$

$$\vdots \qquad \vdots \qquad \vdots \qquad \vdots$$

$$\frac{m_2^P}{m_1^P}, \frac{m_3^P}{m_1^P}, \frac{m_4^P}{m_1^P} \cdots\cdots \frac{m_C^P}{m_1^P}$$

That is, the total number of intensive variables is $2+P(C-1)$.

The independent equations which have to be satisfied by the system in equilibrium are as follows:

$$\mu_1' = \mu_1'' = \mu_1''' = \cdots\cdots \mu_1^P$$
$$\mu_2' = \mu_2'' = \mu_2''' = \cdots\cdots \mu_2^P$$
$$\vdots \qquad \vdots \qquad \vdots \qquad \vdots$$
$$\mu_C' = \mu_C'' = \mu_C''' = \cdots\cdots \mu_C^P$$

That is, there are $C(P-1)$ equations which must be satisfied. The number of variables which are left indeterminate are the number of variables in excess of the number of equations. Therefore, the number of degrees of freedom F, which is the number of indeterminate intensive variables, is shown by the equation

$$F=2+P(C-1)-C(P-1)$$
$$\text{or,} \qquad F=2-P+C$$

The last expression is the form generally encountered in the literature.

In this derivation it was assumed that each of the C components was found in every one of the P phases. However, if a component is absent from some phase, the same rule will be obtained because the ratio of the mass of that element in that phase will be absent, thus reducing the intensive variables by one, and the chemical potential for that component in the given phase will be absent, thus reducing the number of equations by one.

In metallurgical work, however, one of the intensive variables, pressure, is usually constant (atmospheric). This obviously reduces the number of intensive variables by one but makes no change in the number of equations. The phase rule then becomes

$$F=1-P+C$$

This is the form of the phase rule which serves as a criterion when mapping the temperature-concentration equilibrium diagrams customarily used in metallurgical work.*

It is particularly important to check the number of phases present along reaction horizontals representing invariant equilibria. Diagrams have been published showing four phases along the reaction horizontal in a binary system. Application of the phase rule would have prevented such an error.

The phase rule gives no information concerning the location or the shape of the phase boundary. There are two thermodynamic equations which do give some information on this subject:

$$\log_e x' = -\frac{L}{RT} + C \qquad (1)$$

$$\log_e\left(\frac{x_1}{x_s}\right) = \frac{L}{R}\left(\frac{1}{T_0}-\frac{1}{T}\right) \qquad (2)$$

where x' = mol fraction of alloying element
x_1 = mol fraction of solvent in liquid phase
x_s = mol fraction of solvent in solid phase
L = molal heat of solution
R = gas constant
T = temperature, degrees absolute
T_0 = freezing point of solvent, degrees absolute
C = integration constant

Equation (1), which was first developed by LeChatelier, gives the relation between the concentration of a solution and the temperature when the precipitating phase is a pure element or compound. Equation (2), which was first developed by Van Laar, gives the same relation when the precipitating phase is a solid solution. In the derivation of both these equations the assumptions are made that the solutions are perfect (that is, the activity is equal to the mol fraction) and that the heat of solution remains constant over the temperature range considered. Although these assumptions cannot be expected to be valid in all cases, the equations have been very useful in correlating the equilibrium relations in binary aluminum alloys of high purity.

Various rules for the shape of certain portions of phase boundaries have been derived from thermodynamic considerations but will not be discussed here. In this regard the reader is referred to an article by Wilson in *Journal* of the Institute of Metals, V. 70, 1944, p. 543.

*In case the singular pressure were one atmosphere this condensed form of the equation would not apply, but no such case is known for metal systems.

Charles Ulysses Scott

Dean of

American

Heat Treaters

"IN THE PHENOMENAL YEAR OF 1870," the burly old steel treater boasted, with a shrewd twinkle, "the Franco-Prussian War began, Italy became a kingdom, Robert E. Lee died, but Aguinaldo and I — Charles Ulysses Scott — were born." This makes large-speaking C. U. Scott at 78 the dean of steel treaters in the U.S. and therefore someone to contemplate. He began when the blacksmith forge and quenching barrel made up the sole heat treating equipment in a community. He began when the "expert" heat treater was in the toolroom and had a formula of secret ingredients which he mixed at home and carried to work in a paper bag. Scott's first job as a 16-year-old was swinging a man-size sledge, punching eyes in hammer blanks for the Hardsoegs who then, as they still do, made coal and rock drilling equipment at Ottumwa, in Iowa's coal and manufacturing country. This was near Albia where he was born.

He was 22 when he started his own business, a blacksmith shop, in Ottumwa. His specialty was making aluminum horseshoes for thoroughbreds. He put steel studs in the shoes for ground grippers. Dan Patch, the mighty pacer, won many a race wearing Scott shoes. Scott-shod also was the big black stallion that carried "General" Coxey, that onetime rolling mill hand, when he led his "army" of jobless to President Grover Cleveland's Washington. "They certainly knew how to live off the country," Horseshoer Scott remembers.

He turned his hand, a huge delicate hand, to nursing when the Spanish-American War sent him to Florida as a volunteer and most of his company contracted typhoid. That medical experience made him want to become a surgeon, like one of his brothers, so he went to Chicago, got a job hardening tools in Crane Co.'s machine shop, studied medicine nights. "I also earned some money by making scalpels and other surgical tools; and razors you could use ten times without stropping — if your face could stand it." Overwork forced him to give up medicine.

This restless, ambitious man, now 34, then got a job with the Rock Island Arsenal as a tool hardener (a job he evidently filled with lasting satisfaction, for during World War I he returned as consultant, to supervise the department). But eventually he set up business as "C. U. Scott, at the Head of Wall Street, Davenport, Iowa". As "C. U. Scott & Son" he now (and since 1906) has his establishment in Rock Island, from where he declares it to be the first custom steel treating plant established in the U.S.A.

He continues to live across the Mississippi in Davenport, in a folksy little house which he bought after he had remarried. The present Mrs. Scott, a small woman whom he could lift sitting in one of his hands, was a Davenport milliner to whom he had apparently paid no attention, as they both like to recount, but he says, "It wasn't so sudden. I needed a wife. I looked around. She fit."

His older grandson is now in the family business, and the younger one wants to get in right away without finishing college. "But I'm not letting him," says their grandfather and mentor. (Their father, his only child, died eight years ago.) "They might as well learn the business while they are young, but there's a lot to college degrees in metallurgy and in business science. So the youngster can finish his schooling."

C. U. Scott was a member of the American Steel Treaters Society and the 1918-1919 *Journal* contains the text of two brief talks given by him, one on "Heat Treatment of Drop Hammer Dies" and the other on "The Heat Treatment of High Speed Steel". In the latter he said, "It is the duty of this society to show that the treatment of steel is not a mystery, but an art." This indeed would be the conclusion any observant visitor would form today after inspection of Scott's busy establishment, well equipped to give almost any sort of tool or machine part almost any kind of heat treatment. Commercial heat treaters get only the hard jobs, the pieces which the manufacturer either is not bold enough to handle in his own plant, or which are needed in such small number that economical shop routine cannot be set up. Here is where the resourceful artisan shines.

Much of his early success as a commercial heat treater came from his ability to anneal the hard-to-anneal steels in amazingly short times by methods later known in the scientific metallurgist's vocabulary as "isothermal transformation". However, as his bright particular achievement, Scott regards his process for surface hardening of the 18-8 stainless steels, which is done with such precision that customers come to him from as far as both oceans. He, approaching the end of his long and active career, has considered several schemes whereby he could license either the "secret" method, or establish the sale of the necessary salt bath, but has never been convinced that any arrangement would not lead to widespread piracy. Here is one of those situations where the patent law is inadequate to protect the inventor — indeed defeats the aim which Scott has held since that talk given before the Steel Treaters 30 years ago.

"I ought to retire completely," says 78-year-old Charles Ulysses Scott. "I will when both grandsons get their teeth into this business. Meanwhile, I'm taking it a little bit easier and enjoying myself a lot more. Yes, I'm an old steel treater. I guess there's no more like me. Hobbies? I like good workmanship, good living, good hunting, good clothes, good women." MYRON WEISS

Russian Scientists Criticize Einstein

The Letter*

IN RECENT YEARS the venerable scientist Albert Einstein several times raised his voice against Hitlerite barbarians and against danger of new war, in defense of stable peace and against dreams of militarists to subjugate American science completely. However, in his latest statements he has joined the motley company who believes "world government" may be a panacea for securing world peace, alleging that state sovereignty in the atom age is an old, outmoded idea.

The idea of "world government" reflects the fact that capitalist monopolies need world markets, world sources of raw materials and regions for investment capital. We know this perfectly from our country's recent past. Money received from foreign banks of the Tsarist government suppressed the revolutionary movement in this country and handicapped the development of Russian science and culture. The great October Socialist revolution smashed these chains of economic and political dependence and made our country a truly free and independent state, advanced, with an unprecedented tempo our socialist economy, technique, science and culture, and made our country a reliable bulwark of peace and security of the people.

And now the advocates of "world superstate" invite us to surrender this independence voluntarily in the name of a certain "world government" which with a high sounding trademark conceals world domination of capitalist monopolies.

Einstein proposes a "world parliament" which would have the right to pass final decisions, throwing aside the principle of unanimity of great powers. As is known, the American delegation at United Nations takes advantage of the fact that the vast majority of nations are in dependent position and are forced to adapt their foreign policy to the demands of Washington. There is thus created a mechanical majority which votes on orders from its real bosses. Yet Einstein has gone so far as to say that if the Soviet Union refuses to join such an organization, organized along new lines, other states will have full right to enter this road without the Soviet Union.

These statements actually amount to this: If one does not succeed in turning the United Nations completely into a weapon for imperialistic plans and schemes of the State Department of the United States, it is necessary to blow up this organization from inside and create in its place a new "international" organization without the participation of the Soviet Union and countries of new democracy.

*Extracts from letter published in *The New Times*, Moscow English language paper, and signed by Sergei Vavilov, president of the Academy of Sciences of the U.S.S.R., and three other Russian scientists of international repute.

The Reply†

YOUR ATTITUDE on the significance of the antagonism between socialism and capitalism seems to dominate completely your views on international problems. I share your view that a socialist economy possesses advantages; however, we should not blame capitalism — or, we should say, the system of free enterprise — for all existing social and political evils.

For many reasons the United States is now compelled to emphasize her export trade, and these exports must be financed through loans granted to foreign countries. It is difficult to imagine how these loans will ever be repaid. For all practical purposes, therefore, these loans must be considered gifts. There is a real danger that these gifts may be used as weapons of power politics.

Is it not true, however, that we have stumbled into a state of international affairs which tends to make every invention of our minds and every material good into a weapon, and consequently into a danger for mankind?

Although your letter attacks me for my support of "world government", it is mainly an attack upon the nonsocialistic countries, particularly the United States. I believe that behind the aggressive front there lies a defensive mental attitude which is nothing else but a trend toward an almost unlimited isolationism. However understandable this desire for isolation may be when one realizes what Russia has suffered at the hands of foreign countries, it remains no less disastrous to Russia and to all other nations.

If we hold fast to the concept and practice of unlimited sovereignty, each country reserves the right for itself of pursuing its objectives through warlike means. Under these circumstances, every nation must be prepared for that possibility; it must try with all its might to be superior to anyone else. This objective will dominate more and more our entire public life and will poison our youth long before the catastrophe is itself actually upon us. We must not tolerate this, however, as long as we still retain a tiny bit of calm reasoning and human feelings.

This alone is on my mind in supporting the idea of "world government", without regard to what other people may have in mind when working for the same objective. I advocate world government because I am convinced that there is no other possible way of eliminating the most terrible danger in which man has ever found himself. The objective of avoiding total destruction must have priority over any other objective.

†Extracts from "A Reply to the Soviet Scientists", by Albert Einstein. *Bulletin* of the Atomic Scientists, February 1948.

Presentation of verbatim extracts from important contemporary documents concerning atomic energy does not imply that the Editor agrees with the opinions quoted, nor that they are expressions of A.S.M. policy

Successful artillery ammunition was made of nonaging 0.10% C steel strip, quenched from within the Ac_1-Ac_3 range, not tempered nor stress relieved after forming into cartridge cases. Close thickness tolerances and very smooth nondecarburized surfaces are also essential specifications for the strip.

Prewar French Experience With Steel Cartridge Cases

I RECEIVED A PHOTOSTAT from a friend of mine of the "Notes About Some Steel Cartridge Cases Made in Germany", printed in *Metal Progress* for March 1945. Since much of the information contained in this article had to do with German manufacture prior to the war, and since I had some experience in the same type of operation between January 1939 and May 1940 in various arsenals and cartridge factories in France, advising on proper methods of fabrication of steel made in our own works, it may be well to add a few notes to those already published in America. What follows deals with artillery ammunition for small and medium-sized field guns of 20, 25, 37, 47 and 75-mm. caliber.

Choice of Steel — The first work with which I had contact was for ammunition for a 37-mm. gun mounted on a submarine. I went on the assumption that a nonaging steel should be especially suited for this manufacture, since the work of drawing should induce no brittleness whatever in such a steel. The first ammunition was therefore made of a very soft grade of openhearth steel, killed with an excess of aluminum and titanium. Its tensile strength was 51,000 psi. with elongation of 30% or more. This lot of cartridges proved unsuccessful due principally to the fact that they stuck in the gun chamber and were hard to eject from the breech after firing; briefly, they were "stickers". For this reason the next group of cartridges was made from harder grades of steel, containing 0.08 to 0.12% carbon. (We later found that for small cartridges for infantry and automatic rifles it is better to work with even harder steel containing 0.15 to 0.20% carbon.)

We made this second batch of steel very carefully, killing it with aluminum and titanium, as before, to insure a nonaging quality. We also added a slight amount of chromium (0.15%). The steel was very sound; it contained no long inclusions. The transverse impact values were high, and were only slightly affected by strain-aging — a reduction of 10 to 15% being the maximum shown on the tests. In the formulation of this steel it was our idea that the chromium and titanium content would stabilize the carbides and nitrides, thus counteracting to a considerable extent the disadvantages of aluminum, such as its effect in promoting decarburization during annealing, and even graphitization. (This steel was similar to those studied in the United States by Herty, Hayes, Griffiths and others.)

Proof firing indicated that cartridges with a coarse crystalline grain at the base have a low yield point (28,000 psi. or less) and stick in the breech of the gun. Such cartridges tended to crack if their surface was decarburized. In order to insure the necessary fine grain size, even after critical cold work (10 to 20%), which occurs in fabrication near the base of the case, all annealing after drawing was done at 600 to 650° C. (1110 to 1200° F.). Partial softening of the mouth for

By E. Herzog
Supt. Research Department
Pompey Steel Works, Pompey, France

tapering posed a special problem. We found that necessary uniform conditions could not be secured by passing the ends of the cartridges through a slot heated with gas flames. Coreless induction heating by high-frequency electrical current produced the uniform grain size necessary for this final operation.

The steel works delivered strips or plates to the arsenal, and disks were punched from them. The steel plant delivered material with fully normalized structure for artillery ammunition. Spheroidization had to be especially complete in the higher carbon grades for rifle ammunition (0.15 to 0.20% carbon) which requires very smooth surfaces, and the specifications were unusually strict as to thickness of the strip, the limits being plus or minus 0.05 mm. (0.002 in.). Specially inspected operations in the steel plant started with assembly of raw materials and their refining; hot rolled products were pickled, reduced by cold rolling to 60 or 70% of their original thickness, and the strip then spheroidized by low-temperature anneal. The spheroidizing heat treatment is long and expensive; later, normalized strip was successfully used, but extreme care is necessary to retain the smooth surface of the cold rolled strip.

Much attention was given to the troublesome sticking of steel cartridge cases in gun breeches. I was rather amazed to find that steel requires a tensile strength at least 14,000 psi. higher than brass if it is to be extracted readily after firing. Steel cartridges for small weapons (7.5 or 7.8-mm. caliber) required material in the wall near the cartridge base having a tensile strength of 85,000 psi. Compare this with the strength of brass in a standard cartridge — 65,000 to 70,000 psi., $\frac{3}{8}$ in. above the base. We finally determined that for artillery the required tensile strength of steel cartridges is a function of the explosive pressure as measured by the standard crushing test. This relationship is indicated by the following tabulation:

Explosive Pressure Kg. per Sq.Cm.	Psi.	Required Tensile Strength of Steel Cartridge Wall, ¾ In. Above Base
2600	36,000	85,000 psi.
2800	40,000	92,500
3500	50,000	100,000
4000	57,000	107,000

(The crushing test is more or less standardized for ballistic work. The pressure developed by the explosion of the powder in a small chamber is transmitted by a piston to a small copper cylinder held in a fixture screwed into the chamber's wall. The amount of plastic deformation so produced in the copper cylinder is compared with the stress-strain curve of a similar copper cylinder in static compression, and the forces for equal deformation are considered equal. This test is not supposed to measure effective pressures during the explosion — for which more elaborate electronic equipment has been devised — but is an old test that gives comparative relationships fairly satisfactorily.)

We discovered in our early work with the 37-mm. submarine gun cartridges (marine model 1902), made of 0.06% carbon steel, that stress-relieving of the finished cartridges at 650 to 750° F. was ruinous. Proof firing of cartridges not stress-relieved gave very good results; after stress-relieving 90% were stickers. Stress-relieving only slightly lowered the tensile strength (perhaps 7000 psi. on the average) but a sharp yield point reappeared on autographic tensile diagrams after annealing. (Figure 1 represents the diagrams for microtensile tests made on the Chevenard machine and Fig. 2 the test pieces used.) Table I, p. 679, illustrates these points.

It might be mentioned in passing that the German cartridges which I have examined did not have a very high tensile strength. For example, tests on the 105-mm. type gave from 85,000 to 92,500 psi. (about 10,000 psi. lower than the range quoted for 50-mm. caliber ammunition in the

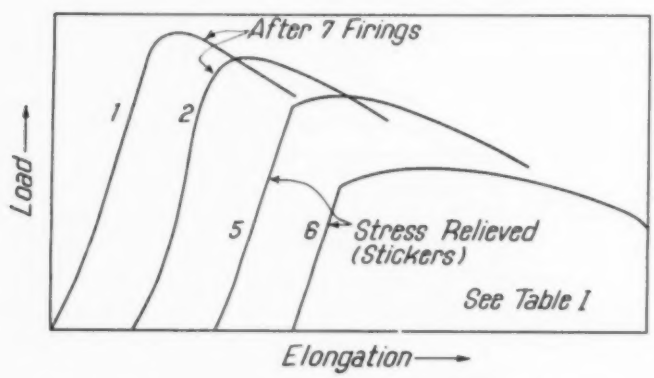
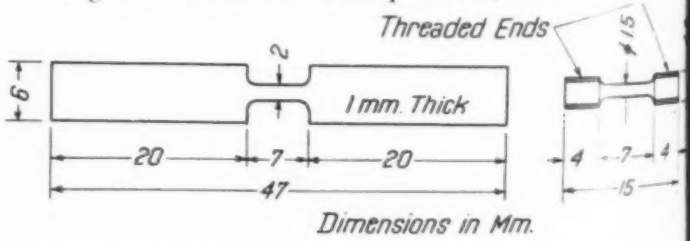


Fig. 1 — Representative Load-Elongation Curves From Autographic Chevenard Micromachine of Steel in 37-Mm. Cartridge Cases; No. 1 and 2 From Cold Drawn Cases After Seven Firings at 37,000-Psi. Service Explosive Pressure; No. 5 and 6 From Unsatisfactory Cases After Stress Relief. See Table I

Fig. 2 — Microtensile Test Specimens, Flat and Round



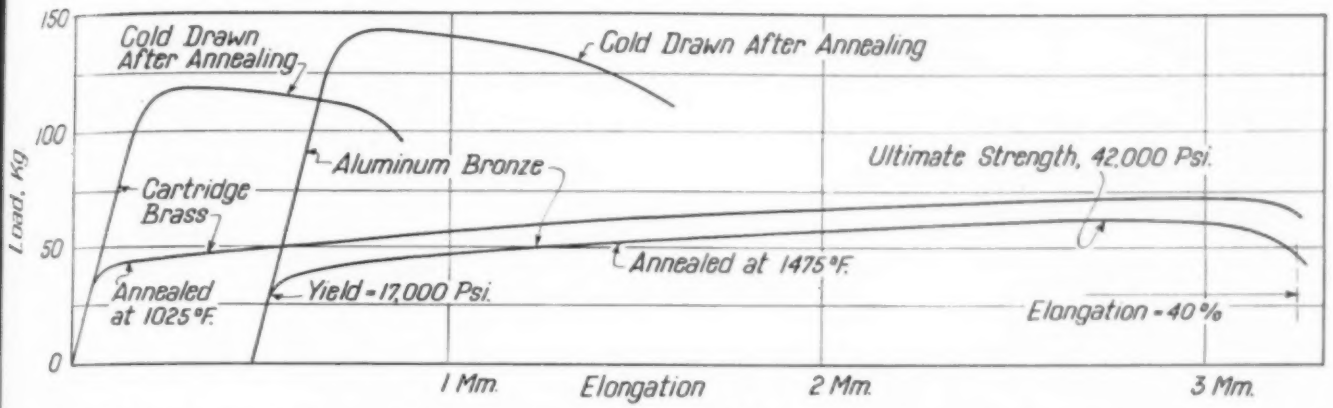


Fig. 3 — Load-Elongation Diagrams for Microtensile Specimens of Cartridge Brass and Aluminum Bronze Showing the "Spring Properties" (High Proportional Limit) Acquired by Cold Work. Principal properties of all four conditions are:

	Yield	Ultimate	Elongation
Brass, annealed	21,000	48,000	47
Cold drawn	60,000	83,000	12
Bronze, annealed	17,000	42,000	40
Cold drawn	71,000	103,000	15

above-mentioned article in *Metal Progress*). The tests given in Table I for the cases which withstood successfully seven firings at 36,000 psi. explosive pressure without malfunctioning are also at considerable variance with the opinion reached by American experts (as quoted in the same article, wherein steel cases with a yield of 85,000 to 90,000 are said to cause a prohibitive number of "stickers", whereas cases with yield strength of 115,000 psi. extracted easily). It would appear that the practice of using unduly strong steel was successfully followed in the manufacture of 3-in. caliber steel cartridges by Norris Stamping & Mfg. Co., of Los Angeles, as described in the January 1945 issue of *Metal Progress* by Fred M. Arnold, although in his article the properties of the steel are graded by hardness impressions rather than by tensile tests.

The high tensile properties of the Norris production are obtained by heat treatment; my experience led to a similar conclusion as to the proper method of getting the required strength for explosive pressures higher than 40,000 psi., and I suggested as early as March 1940 that the cases should be hardened sufficiently to give 115 to 140 psi. tensile strength on cartridges made of soft steel containing 0.08 to 0.12% carbon. Reports of this work, together with a set of cartridges and steel specimens, were forwarded to the British Scientific Liaison Officer in France; whether any of our allies made use of this information I do not know. After liberation the Pompey Steel Works was granted a patent (January 1945) for this strengthening process.

I was led to propose this strengthening process by some general ideas to the effect that steel cartridges should behave like springs; a reversible deformation or recovery after explosion is absolutely necessary. Ordinary cartridge brass (70:30 Cu:Zn) and aluminum bronze (90:10 Cu:Al) behave like springs in the cold drawn state. Even after firing, these cartridges become harder and are strengthened by as much as 7000 psi. or even more in the wall of a 25-mm. cartridge case, ¾ in. above the base.

This general property is exhibited in stress-strain curves of these alloys when tested in the annealed state and after cold drawing about 30%, as is plainly shown in Fig. 3, above.

On the other hand, soft steels in the annealed or normalized condition do not possess a similar large capacity for strength-

Table I — Representative Microtensile Tests From Steel Cases

No.	Location	Elastic Limit	Ultimate Strength	Elongation
	Cases Not Stress-Relieved, After Seven Firings	Psi.	Psi.	
1	⅜ in. above base, transverse, exterior fiber	87,000	101,000	10.5%
2	⅜ in. above base, longitudinal, interior fiber	79,600	92,500	10
3	⅜ in. above base, longitudinal, exterior fiber	71,200	88,300	8.5
4	⅜ in. above base, transverse, interior fiber	58,400	94,000	10
	Cases Stress-Relieved for 10 Min. at 750° F. (Stickers)			
5	⅜ in. above base, longitudinal, exterior fiber	76,000	82,500	11
6	2 in. above base, longitudinal, exterior fiber	57,000	71,400	14
7	3½ in. above base, longitudinal, exterior fiber	72,500	79,600	16
8	Neck	54,000	61,100	14

ening after plastic deformation. Only the austenitic alloys cold work materially — for example, the 18-8 stainless steels and the 25% nickel-iron alloys. Is is noteworthy that these alloys which are easily strengthened by plastic flow have body-centered cubic crystalline lattices. On the other hand, the ferritic and pearlitic steels have less closely packed atoms, the structure being face-centered cubic with lattice dimension of 3.86 Å units, in contrast with 2.86 Å units for the crystals in beta brass and gamma (austenitic) iron.

ture could be drawn into cartridge cases without any special difficulty. (Flat bending is possible with such quenched steel.) Near the base of the cartridge the tensile strength was very high — 115,000 psi. in a region which had suffered 20% reduction of area.

If the steel after having been given the high (Ac_3) quench is tempered at about 1100° F., strengthening by cold rolling is not nearly so pronounced. Yield points of such steel after corresponding cold reduction are about 14,000 psi. on

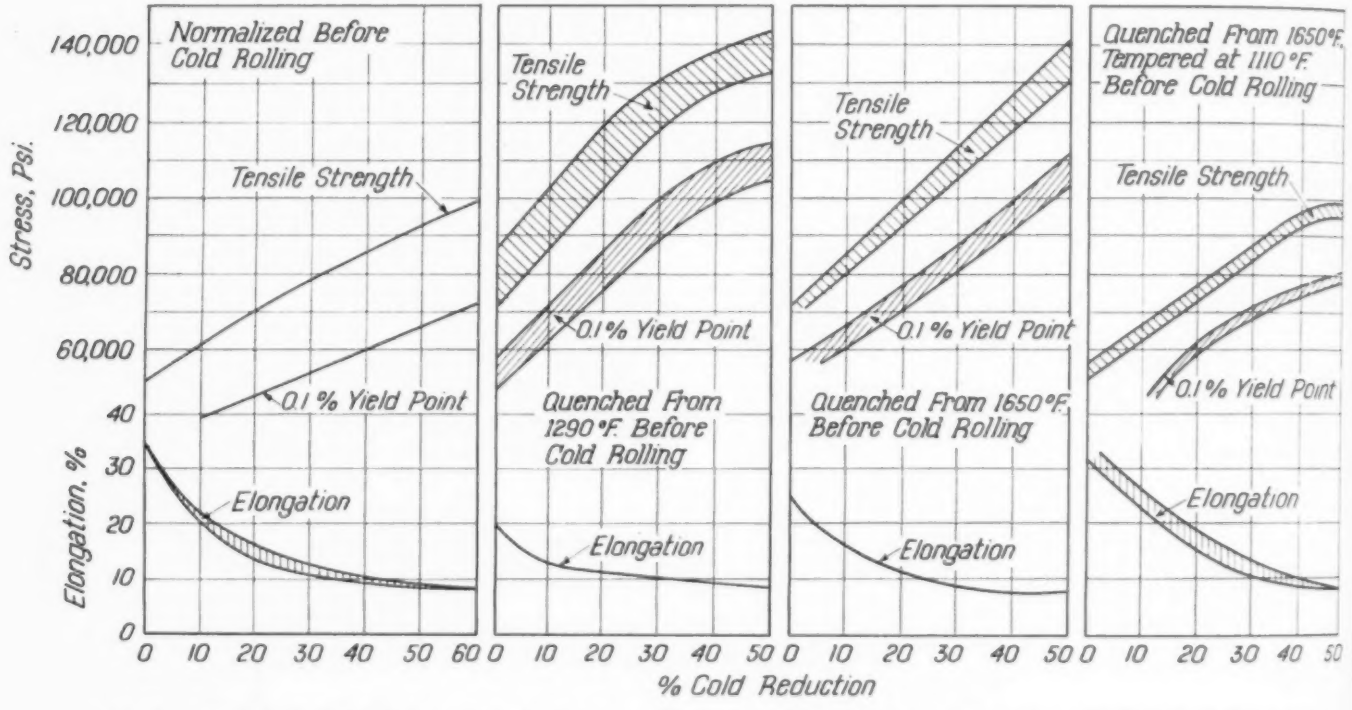

Normalized Before Cold Rolling
Tensile Strength
0.1 % Yield Point
Elongation
Quenched From 1290 °F. Before Cold Rolling
Quenched From 1650 °F. Before Cold Rolling
Quenched From 1650 °F. Tempered at 1110 °F. Before Cold Rolling
Stress, Psi.
Elongation, %
% Cold Reduction

Fig. 4 — Result of Cold Work on a Killed Openhearth Steel (0.08% C, 0.43% Mn, 0.04% Si) of Nonaging Characteristics in Normalized, Low-Quenched, High-Quenched, and High-Quenched and Tempered Condition

It therefore seemed to me to be logical to test the effect of plastic flow on quenched soft steels. Results were very good, as shown in Fig. 4. These diagrams show the effect of cold rolling a killed openhearth steel containing 0.08% carbon, 0.43% manganese, and 0.04% silicon. The steel was of a nonaging type, and its original condition was in hot rolled plates, 4 mm. thick, normalized. These plates were cold rolled various percentages up to 60%, and microtensile specimens, 1.5 mm. in diameter (Fig. 2), were produced suitable for testing in the Chevenard machine.

It will be observed that quenching at the Ac_1 point — that is, 1300 to 1380° F. — is more efficient than quenching from the Ac_3 point (1560 to 1650° F.). I found that disks made from this steel quenched from 1300° F. in water at room temperature could be drawn into cartridge cases without any special difficulty. (Flat bending is possible with such quenched steel.) Near the base of the cartridge the tensile strength was very high — 115,000 psi. in a region which had suffered 20% reduction of area.

If the steel after having been given the high (Ac_3) quench is tempered at about 1100° F., strengthening by cold rolling is not nearly so pronounced. Yield points of such steel after corresponding cold reduction are about 14,000 psi. on the average higher than if the steel were cold drawn in the normalized condition.

The explanation of this relation between hardening by quenching and cold work is rather obscure. It would appear that several factors are superimposed, but it seems reasonable to assume that a solid solution of carbon and other elements is retained in the alpha iron from the solid solution existing in gamma iron just prior to quenching, and this alpha solid solution is decomposed by the cold work. Micrographs of material quenched from 1380° F. (that is, between Ac_1 and Ac_3) show a two-phased structure — about half ferrite and half pearlite in a 0.12% carbon steel. We believe that the ferrite phase contains about 0.03% carbon in solid solution and the fine pearlite represents the transformed austenite existing at

the quenching temperature. It should analyze about 0.21% carbon, on the basis of an equal partition of the quenched structure between ferrite and fine pearlite.

Figure 5 shows this structure after cold rolling 50%; the microstructure does not give any very clear reason for the enhanced hardening during cold working, but of course there may be submicroscopic phenomena involved.

It remains to be explained why the German steel cartridges withstood high explosive pressures, as in anti-aircraft and anti-tank guns, yet had a fairly low yield strength. I believe that this is due to the method of protection against corrosion, which consisted of a thin layer of lacquer or other organic matter. The gases evolved when this coating was heated from the powder explosion effectually prevented sticking in the breech. This solution of the problem of stickers may not be best, for if the gun becomes fouled by decomposed carbonaceous matter, the cartridges will not eject properly. Another point to which the Germans doubtless paid attention was the design of the gun chamber. For example, extraction of a cylindrical cartridge will always be very much more difficult than one which has a slight conical taper. This point was touched upon in Mr. Arnold's article printed in *Metal Progress* for January 1945.

Fig. 5 — Steel of Fig. 4, Quenched From 1290° F., Reduced 50% by Cold Rolling. 700×

Critical Points

By the Editor

To ALLIS-CHALMERS to talk about metallurgy of gas turbine parts and found the staff deep in the problems of designing a locomotive power plant using powdered coal as fuel — this in conjunction with the Bituminous Coal Research, Inc. Allis-Chalmers has, probably, the longest American experience with gas turbines, having (since 1938) built 190,000 hp. of gas turbines for driving centrifugal air compressors in Houdry process synthetic gasoline plants. These turbines operate at relatively low temperature (blades at about 900° F.) and at correspondingly low thermal efficiency, a matter of no great moment when a surplus of byproduct gas is available. Also, metallurgical problems due to heat are no worse than in a modern steam turbine, except perhaps in the flame tubes. At any rate, this long experience was available when designing and constructing the large experimental gas turbine now undergoing a long series of tests at the Naval Engineering Experiment Station in Annapolis. Designed for top temperature of 1300° F. at the blades, this turbine has already run several 8-hr. trials with gas at 1500° F. without observable damage. Principal problem in the locomotive design, other than space limitations, is the effect of fine ash from the powdered coal. The Bituminous Coal laboratory is making a special study of the relationship of such things as velocity, turbulence and angle of impingement to scour and corrosion by fly ash in the gas. Under certain circumstances ash particles adhere to the hot metal to form a self-renewing coating of viscous slag that

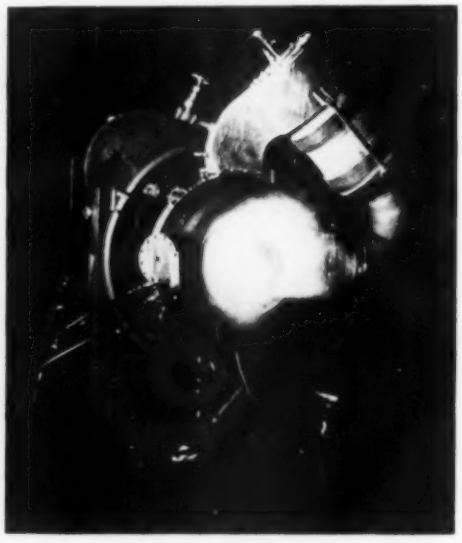
Gas turbines for peaceful industry

acts as a protection for the underlying metal, thus giving, in effect, the ceramic sought by so many investigators.

DESIGN, constructional and metallurgical problems in these heavy and bulky engines of peace are obviously on a different order from those of aircraft turbojets, where weight, size, and thermal efficiency are paramount considerations. Since efficiency increases with working gas temperature, there is practically no end to the designer's demands for rotor and nozzle blades to operate at higher and higher temperature. These, being small, are easier to cast of special alloys than it is to make the larger disk itself. Consider this combination of required properties before you proceed to design a new alloy for turbine disks: Yield strength (0.2% offset) at 70° F. more than 80,000 psi., coupled with at least 10% elongation, rupture strength more than 40,000 psi. for 1000 hr. at 1200° F., and with at least 5% elongation at rupture. The alloy must be forgeable, machinable, weldable, and resistant to heat shock. It should require but small proportions of strategic and critical metals like columbium. If it can be melted and cast commercially, without using vacuum equipment or unusually inert atmospheres, it would be a mark in its favor.

Requirements of disks

Courtesy General Electric Co.



Test on a Turbojet

BLADES and buckets for turbojet aircraft engines are usually rather small affairs, weighing in ounces, and so the investment molding practice used by dental laboratories for precision casting of dentures has been successfully adapted to the unforgeable, unmachinable alloys high in cobalt. This is an old story, but the Allis-Chalmers' precision casting department contains enough labor-saving equipment so it looks more like a production foundry than a dental laboratory 1000 times multiplied. Since this is a production operation, Kenneth Geist, engineer in charge, is out for business and believes his castings can compete with many forgings of stainless steel because the forging dies are rapidly cut by abrasive scale. Small, intricate levers for sewing and business machines have been cast in alloys suitable for nitriding; also small parts of 8620, later to be carburized. A field under active investigation is complicated shapers and milling cutters of tool alloys. So far the heaviest investment casting made by Allis-Chalmers weighs about 10 lb.

Precision castings in mass production

TO RETURN to high-temperature alloys, a tough metallurgical problem in gas turbines is the flame tube of combustion chambers. Pure ductile molybdenum sheet — now that this is available from methods developed in Climax Molybdenum Co.'s laboratories — may satisfy the requirements of formability, resistance to warpage, and refractoriness up to 1900 or 2000° F. Molybdenum melts at about 4750° F., but unfortunately it oxidizes readily and the oxide is a volatile, voluminous white cloud that seems to attack surrounding metal . . . . At a meeting of the Columbus Chapter — national officers' night, with the Editor attempting to pinch-hit for Bill Eisenman, but failing to reach first base — was talking with President Foley, Howard Cross of Battelle Memorial Institute, and Mars Fontana of Ohio State University, and was told about a curious phenomenon called "catastrophic oxidation" by Foley. Many high-alloy steels, with molybdenum of at least 6%, will revert to a pile of oxide overnight in stagnant furnace atmosphere at 1300° F. or more, but be unaffected by moving gas at the same temperature. For example, a flat bar on a rough hearth will start catastrophic oxidation *from the bottom;* if it is on a smooth bar of heat resistant material it is immune (very little oxygen reaches it); if it is blocked up on refractory it is immune also (gas circulation below). Apparently, a moving atmosphere carries away the small amount of molybdic oxide that forms at the steel's surface; if it remains there it reacts with the other metal atoms in contact, transferring to them some exceedingly active atomic oxygen, and then turning right around and picking up some more from the surrounding atmosphere. (Maybe that's not too precise a description of a catalytic effect.) Indications are that tungsten oxide may also cause this destructive effect.

Catastrophic oxidation of molybdenum

REMINISCING with R. J. Cowan on furnace developments and got an interesting sidelight on Surface Combustion's prewar development of an atmosphere generator using kerosene, and heard of an unexpected order for 40 of them by the Russian government long before Hitler's invasion, and about the Russians he knew who were certainly long on the theory of heat treating. So got to theorizing ourselves and decided that much was yet to be learned about the iron-carbon-silicon system (I mean cast iron) — especially how to make the solidified metal cool down as a mixture of iron and nodular graphite instead of iron, carbide and plates of graphite (gray iron), or pearlite and cementite (white iron). If this item of "pure science" could be clarified, and if it turned out that a mixture could be properly melted, inoculated and then cast so it was supercooled before solidification, we could get true malleable iron (ferrite plus spherulites of graphite) during cooling, direct from the mold. Such a scientific discovery would certainly revolutionize the American malleable industry.

Cast iron, malleable as cast

IN AN all-too-brief inspection of the Rome works of Revere Copper & Brass, was shown the equipment just getting under way for a new, precise study of machinability of screw stock and other copper alloys. Marvelous in its precision and adorned with electronic equipment for supravisual observation and instantaneous recordings, it is indeed a far advance over the husky lathe with which Taylor and White did their epoch-making work with toolsteel. Their problem, in reverse, is being tackled 50 years later by Harry Schagrin and his men in the metallurgical and manufacturing department of Revere. *Then* it was only necessary to start turning a big round of tough steel, cutting $\frac{3}{16}$ in. deep at $\frac{1}{16}$ in. feed and adjusting the speed until the tool was ruined in 1½ hr. While this scheme of hogging off the metal was wholly competent to differentiate between ordinary carbon toolsteel with 15-ft.-per-min. speed, air hardening tools (30 ft. per min.), and best high speed heat treated properly (100 ft. per min.), it naturally would be entirely inadequate to distinguish between minor variations when lightly cutting the machinable alloys. Consequently, at Revere's Rome plant is installed a special Monarch lathe built to tolerances that are close even for a modern toolmaker, a lathe balanced dynamically without gearing in the headstock, and anchored to concrete reaching deep down to undisturbed foundation. This lathe is driven through a rubber coupling by a motor which is also dynamically balanced and independently anchored. Speed can be varied at will by stepless increments, and then held closely constant by a battery of electronic equipment that passes metallurgical understanding. At present it is thought that machinability can be related to variations in tool-tip pressure (for standardized tool shape and size of chip). Pressure is measured by a compression capsule in the tool post, variations are recorded automatically in figures, tremulous fluctuations shown on an oscilloscope screen and photographed on movie film which notches forward a little in synchronization with the travel of the scanning spot of light. A print from it looks like lamellar pearlite, a trifle nervous.

Machinability measured with precision

VISITING with C. U. Scott in Rock Island, a real veteran in commercial heat treating, and talked long with him about his "secret" process of surface hardening stainless. Secrecy still exists in many commercial operations, and not alone in the shop. In this connection my mind turned back a full year to a pleasant luncheon in San Francisco's Press Club with Kenneth V. King, the energetic materials engineer on Standard Oil Co. of Calif.'s management, who has been so active in the regional affairs of the American Welding Society. The burden of our conversation (other than the salubrity of the region and the friendliness of its inhabitants — topics on which there was perfect accord) was the correct adjustment between the natural urge of research workers to inform their professional brethren about their discoveries — that is, to publish — and the equally natural desire of a business management to retain essential information until its value has been capitalized. Since industrial research is almost exclusively in applied science rather than in fundamental science, something can be said for the businessman's point of view, although it must not extend to the place where the research department is surrounded by a semipermeable membrane which filters information inward but none outward — if for no other reason than that good and desirable researchers will leave such a place as soon as a decent post is found elsewhere.

Danger in secrecy

Those few executives who hold the membrane idea should read and reread the remarks by Albert W. Hull before a recent conference on physical science and human values at Princeton University. The assistant director of General Electric's Research Laboratories said: "In the 30 years I have been with General Electric, I haven't known of a single case where we have lost by publication — or even by telling, in its preliminary stages, what we were doing."

The third of a four-installment article on "The Eight Forms of Corrosion", wherein the entire field of metallic corrosion is broadly considered. Dezincification was one of the earliest forms of selective attack that was intently studied and understood.

Dezincification and Erosion-Corrosion

IN THE February and March issues of *Metal Progress* were printed two articles considering some of the various forms in which corrosive attack on metal manifests itself. The matter now to follow is part of the complete article (intended eventually as an introductory chapter in the new edition of "The Book of Stainless Steels").

At the outset eight forms of corrosion were listed:

1. Uniform attack or general corrosion
2. Intergranular corrosion
3. Pitting
4. Galvanic or two-metal corrosion
5. Concentration cell corrosion
6. Dezincification
7. Erosion-corrosion
8. Stress-corrosion

The first five of the items have been discussed in the previous issues. We will now proceed to say a few words about dezincification and erosion-corrosion.

Dezincification

Dezincification, as the term implies, was first observed on brasses, which are copper-zinc alloys. The zinc is selectively leached out of the brass to leave a weak porous mass of copper. Dezincification can be readily observed, since the attacked areas show the red color of copper as compared to the distinctive yellow of brass.

The mechanism of dezincification is still somewhat controversial, but one commonly accepted envisages the complete solution of the alloy; the zinc stays in solution, but the copper plates back onto the metal. There are two general types of dezincification, and both are readily recognizable. The more common type is uniform; dezincification proceeds at substantially the same rate over the entire exposed area. The second is the so-called plug-type dezincification which occurs in localized areas.

Figure 1 shows a section of brass pipe after service in handling water. The dark inner portion of the pipe is dezincified alloy. The outer portion is the unaffected yellow brass. Dezincification has progressed uniformly through half the wall thickness.

Figure 2 is an excellent example of plug-type dezincification. The dark areas are the dezincified metal with the remaining portions of the pipe unaffected. A cross section of one of the plugs is shown in Fig. 3. Attack was from the inside surface of the tubing in an aftercooler in a plant powerhouse. This heat exchanger carried water inside the tubes, with flue gases at 350° F. outside the tubes. These tubes were made from muntz metal, which is a high-zinc brass, so susceptible to this type of attack that it has been largely supplanted by other copper-zinc alloys.

In general, the dimensions of the brass parts do not change appreciably during dezincification;

By Mars G. Fontana
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Director, Corrosion Research Laboratory
Ohio State University

this may account for the unexpected failures. In one instance a perforated brass plate in a tower handling a warm mixture of water and acetone appeared unattacked after several years of service, except for the coppery color on the surface. Examination showed that the plate was almost completely dezincified, and it cracked completely after slight bending. Brass is a very ductile material. In another heat exchanger, handling boiler feed water, many tubes failed because of plug-type dezincification in only a few areas on each of them.

The uniform or layer type of dezincification occurs most frequently in the high brasses (high-zinc content), and the plug-type seems to favor the low brasses (lower zinc content). More aggressive corrosion conditions — for example, higher temperatures — also favor the plug type of attack, which is related to pitting in that localized attack is involved. Dezincification sometimes occurs under supposedly mild conditions of corrosion, but favorable conditions for this type of attack involve a good electrolyte such as sea water, slightly acid conditions, or the presence of appreciable amounts of oxygen in the liquid handled.

Fig. 1 — Uniform Dezincification Proceeding Half Way Through a Brass Water Pipe

Fig. 2 — Plug-Type Dezincification of Muntz Metal Pipe in an Aftercooler (Boiler Water Inside, Hot Flue Gas Outside)

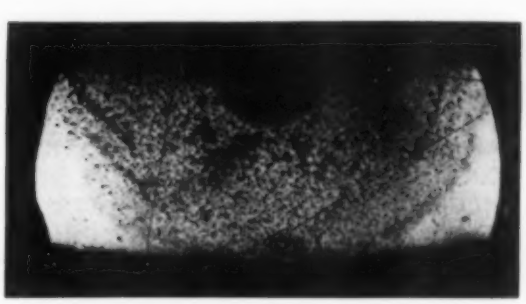

Fig. 3 — Cross Section (Magnified) Through One of the "Plugs" in Fig. 2

The addition of tin, antimony, arsenic, and phosphorus to brass aids in resistance to dezincification. Admiralty metal, for example, contains approximately 1% tin, 29% zinc, and 70% copper. Such alloys, plus definite inhibitors, are often used to replace straight brass where service is accompanied by mild dezincification. For more severe conditions, red brass (15% zinc) is often used because this alloy is practically immune. For very severe conditions it is often economical to use the binary copper-nickel alloys.

The phenomenon of selective leaching of one of the elements of an alloy is by no means confined to the brasses. For example, a hard facing alloy of cobalt, chromium and tungsten known as Stellite No. 1 became soft after several weeks of handling a waste sulphuric acid slurry containing solid ferrous sulphate. Chemical analysis of the soft material and the hard, unaffected alloy showed that cobalt had been selectively removed, leaving a weak porous structure.

Erosion-Corrosion

Erosion-corrosion often causes unexpected and rapid deterioration of plant process equipment. The rate of corrosion can be greatly accelerated when mechanical or abrasive conditions are present, such as liquids moving at substantial velocities, presence of solids in suspension (that is, slurries), marked turbulence, and impingement of fluid streams. Examples of equipment that may be subject to erosion-corrosion are pumps, valves, centrifugals, measuring devices (such as an orifice), agitators, pipe lines, particularly at elbows and tees, and heat exchanger tubes. The corrosion effect known as "cavitation" is one type of erosion-corrosion.

Most metals and alloys depend upon the automatic development of a thin protective surface for their resistance to corrosion. This may consist of an oxide film, a corrosion product (such as lead

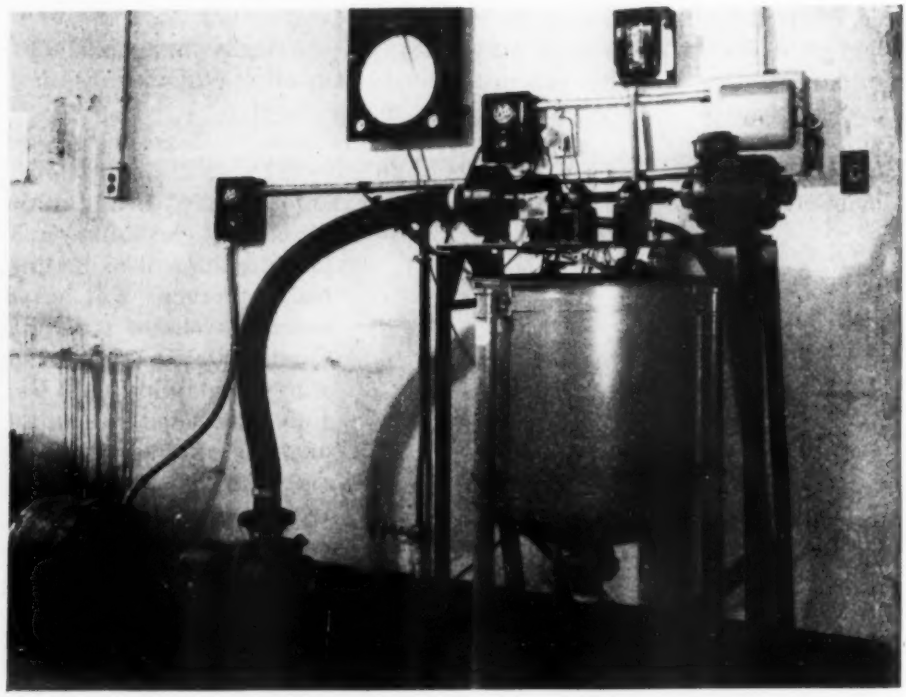

Fig. 4 — Erosion-Corrosion Test Equipment Mounted on Cylindrical Tank Holding Liquid or Slurry to Be Handled. Circulating pump and motor on floor at left

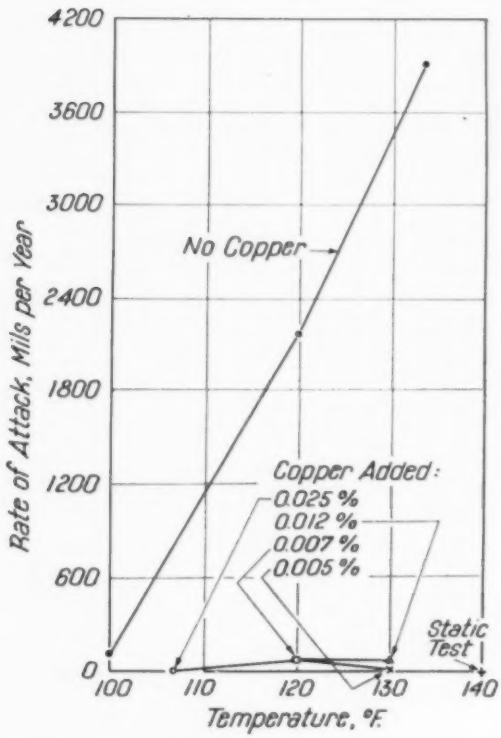

Fig. 6 — Inhibiting Effect of Copper in a Sulphuric Acid Slurry on the Erosion-Corrosion of 18-8 S Mo Stainless Steel

sulphate which forms on lead in sulphuric acid), or an adsorbed film of gas. If protective surfaces are removed locally by abrasion and fresh metal exposed, the material deteriorates more rapidly than under the sole effects of corrosion.

An excellent example of erosion-corrosion involves hard lead valves in hot, dilute sulphuric acid. Coupon tests made of hard lead immersed in this acid showed practically no corrosion, but throttled valves failed in less than one week! In this same service, lead lines with slight bends in them were cut through in a few months.

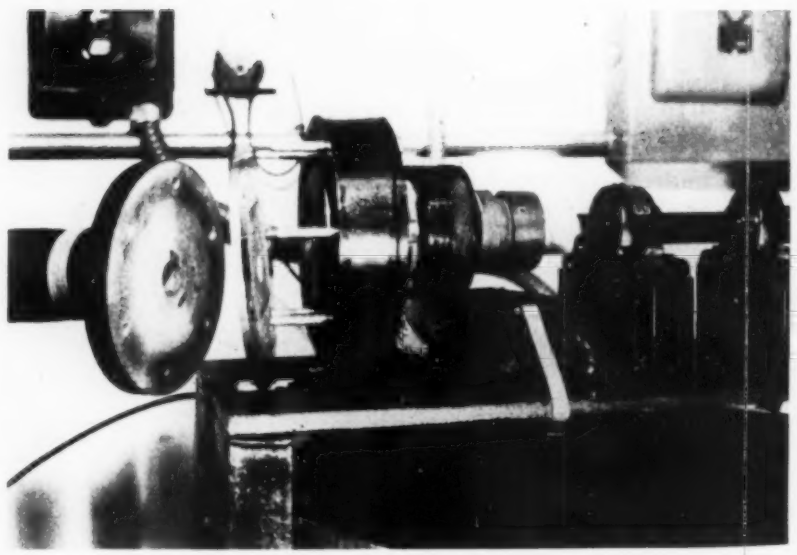

Fig. 5 — Exploded View of Nonmetallic Housing for Erosion-Corrosion Test. In housing at right is the disk specimen. Rotating disk and stationary wear plate (on assembly) are separated by small clearance through which liquid is forced at high velocity

Unfortunately, the usual and common static corrosion test does not include such effects of erosion as high velocity. As a result, equipment often fails within a short time, although the static corrosion tests indicate very long life. For example, static tests, and even tests wherein the specimens were the propellers on a high-speed mixer, showed practically no attack on 18-8 Mo stainless steel in a slurry consisting of waste sulphuric acid and ferrous sulphate. However, the solid bowl centrifuges, built of the same alloy to separate the solids from the acid, deteriorated

in actual service at a rate of approximately 5000 mils (5 in.) per year. This experience clearly demonstrates the need for test equipment that simulates actual operating conditions.

Special Testing Equipment

This problem was solved through the use of erosion-corrosion test equipment, somewhat similar to that shown in Fig. 4 and 5. The results of this study (Fig. 6) indicate that no corrosion occurred in the static test, even when the temperature was raised to 140° F. The rate of deterioration under conditions of erosion-corrosion, however, is great and of the same order of magnitude as the rate of attack on the actual centrifugals. The erosion-corrosion test unit thus provided a tool for studying the problem. Figure 6 shows the beneficial effect of adding a slight amount of copper to the acid slurry. Maintenance on the centrifugals was reduced to a normal amount as a result of this study.

The equipment shown in Fig. 4 consists of a 30-gal. glass-lined tank, neoprene hoses, a "Chlorimet 3" pump, a nonmetallic housing for the test specimen, and auxiliary equipment such as heaters and temperature controls. The pump shown on the floor in Fig. 4 recirculates the liquid or slurry contained in the cylindrical tank. The only metallic parts in contact with the solution are the pump and the test specimens. Figure 5 is an exploded view of the housing, the 4-in. test specimen disk, and the wear plate. The disk rotates at 1750 r.p.m. A small and adjustable clearance exists between the disk and wear plate. The liquid is forced past the surface of the disk by the circulating pump and thus obtains the required velocity effect. The duration of a test is usually overnight or about 16 hr.

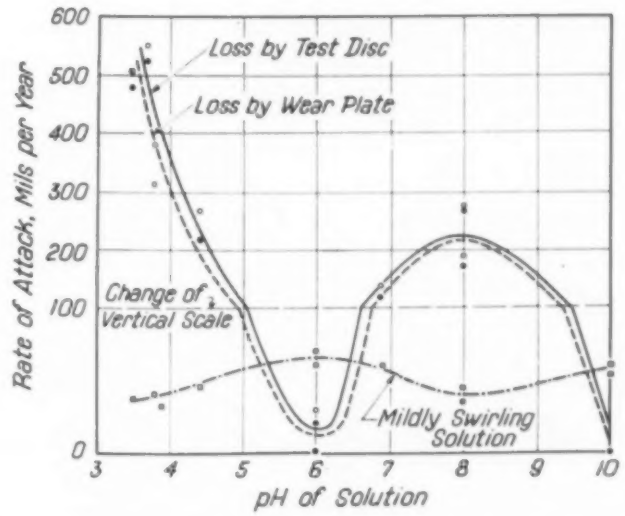

Fig. 7 — Comparison of Erosion-Corrosion Tests of S.A.E. 1020 and in Mild Swirling Exposure to Potable Water at 122° F. 1/16-in. clearance, disk to wear plate

Fig. 8 — Impingement Attack on Elbow in Steam Condensate Return Line

Figure 7 shows data of other erosion-corrosion tests on S.A.E. 1020 steel in water at 122° F. There is a large difference between the erosion-corrosion data and the data obtained on specimens (solution tests) hung in the glass-lined tank, subject only to a mild swirling motion.

Figure 8 shows what is left of an elbow removed from a steam condensate return line. The vertical straight run of pipe just ahead of this elbow showed very little attack, but the elbow failed where the liquid was forced to change direction and impingement effects were present.

Finally, Fig. 9 graphically shows the destructive effect of erosion-corrosion on a high-alloy stainless pump impeller after a relatively few weeks in service. It is obvious that sound engineering must pay attention to velocity of corroding medium as well as its nature.

Fig. 9 — Remains of Stainless Steel Pump Impeller, Ruined by Erosion-Corrosion

Correspondence

Replacing Tungsten With Aluminum in High Speed Steels

LEOBEN, STEIERMARK, AUSTRIA

To the Readers of METAL PROGRESS:

Continual development over the years to reduce the tungsten content in the conventional 18-4-1 high speed steel — 18% tungsten, 4% chromium and 1% vanadium — has led to the adoption or increase of molybdenum, cobalt, chromium or vanadium. Such steels have a hypereutectoid structure in the fully hardened state, in which carbides are embedded in a martensite matrix. Both of these constituents in high speed steels *have and retain* higher hardness at high temperature than their counterparts in plain carbon steel.

In our pursuit of this problem of saving tungsten in Austria (where we have no native tungsten ores) we have adopted a working hypothesis to guide our research. We assume that alloy carbides are formed more slowly than iron carbide, hindering alloy depletion and consequent softening of the matrix and that, in this, the action of tungsten (or molybdenum) is twofold: More atomic species are necessary for the alloy carbides, and the relatively large tungsten and molybdenum atoms create diffusion-impeding distortions in the matrix lattice. (The atomic radii of W, Mo and Fe are 1.41, 1.40 and 1.27 Å, respectively.) It was felt that the latter action could be duplicated by other large atoms, even though the species did not form carbides. It was realized that if such elements were used, a minimum amount of carbide-forming element must be retained.

In the light of this hypothesis, it was felt that aluminum (atomic radius, 1.4 Å) held some promise as an alloying element in high speed steels.

In the course of a general metallographic study of high speed steels it was found that if more than 1.6 to 1.8% vanadium were present, vanadium carbide (VC) existed as a separate phase over and above the normal complex (ledeburite) carbide, $(Cr,Mo,M,V)_mC_n$, and that this critical vanadium percentage was relatively unaffected by wide changes in other alloy content. Furthermore, it was discovered that aluminum did not form carbides or enter into solution with the complex carbides normally present in such steels, but remained in the matrix.

While aluminum appears to effect little change in high-alloy steels such as 18-4-1 high speed, it does increase the resistance to softening at high operating temperatures in low-alloy steels. We explain this difference (in line with our working hypothesis) by saying that sufficient tungsten atoms are present in the matrix of high-alloy steels to impede diffusion through and subsequent softening of the matrix, whereas in low-alloy steels,

Table I — Cutting Ability of Substitute High Speed Steels

	STEEL A	STEEL B	STEEL C	STEEL D	STEEL E
Composition*					
Tungsten	None	None	1.30	2.50	1.30
Aluminum	None	0.50	None	None	0.50
Minimum Tool Life in Minutes					
Test log A					
At 12 m. per min.	22	48	43	—	64
At 13 m. per min.	17	26	25	—	31
At 14 m. per min.	13	15	15	—	17
Test log B					
At 18 m. per min.	21	39		96	
At 19 m. per min.	11	20		42	
At 20 m. per min.	6	12		23	

*Other elements (approximate): 0.90% C, 0.40% Si, 0.40% Mn, 3.0% Mo, 4.0% Cr, 2.5% V.

what little tungsten is present exists in the carbides produced during tempering, allowing the aluminum which remains in the matrix to operate as a diffusion-impeding element.

Interest, therefore, centers upon the low-alloy steels where aluminum is capable of effecting some improvement in quality. For these steels (specifically steels of the basic composition 4% Cr, 2.5 to 3.0% each of W, Mo, and V were studied), it was found that *(a)* the carbon should be greater than 0.85% (to insure the presence of the hard VC) but not greater than 1.1% (since for these carbon contents the complex carbides melted at the necessary quenching temperatures, 2245 to 2280° F., and large amounts of relatively soft austenite were produced); *(b)* the silicon should not exceed 0.5% (otherwise abnormal grain growth occurred); and *(c)* the aluminum should run at least 0.4%, to obtain maximum improvement in cutting properties, but not over 0.6%, to prevent abnormal grain growth and consequent embrittlement of the steel.

These steels with 0.4 to 0.6% aluminum require the same quenching temperatures (2245 to 2280° F.) as their aluminum-free counterparts. The tempering differs in that the aluminum-bearing steels should be tempered twice at 1040° F. for 30 min. to obtain full hardness.

Comparative lathe tests on five of these steels were run at four independent shops. In Table I it is seen that a W-free Cr-Mo-V steel (Steel A) gives poor performance; that the same steel with 0.5% Al (Steel B) gives an improved performance, about equal to the same steel with 1.3% W (Steel C) but not as much as the same steel with 2.5% W (Steel D) or with 1.3% W plus 0.5% Al (Steel E).

On the basis of these tests, Steels B and D were released to the trade. Experience now shows them to equal Cr-Mo-V steel with 2.5% W on lathe work and to have performance comparable to 10% W steel for drills.

As with all low-alloy high speed steels, the best cutting performance is obtained only with optimum heat treatment. Comparison of time-speed curves of tools hardened from 2300° F. and from 2245° F. showed a 50% reduction in performance when quenched from the higher temperature.

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Metallurgical Dept.
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Magnetic Particle Inspection of Silver-Plated or Phenolic Resin Coated Steel

EAST HARTFORD, CONN.

To the Readers of METAL PROGRESS:

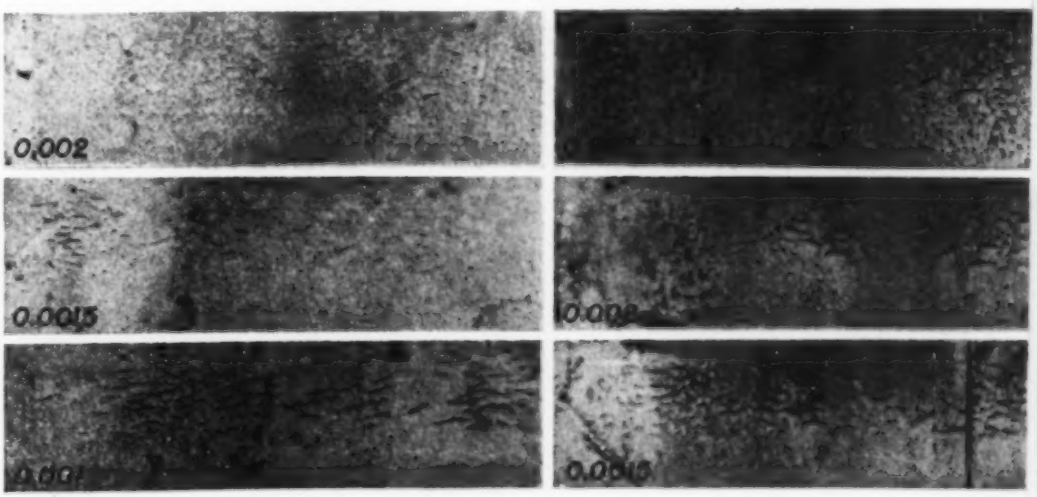
The article "Magnetic Particle Inspection of Chromium Plated Tools" by M. H. Mueller and W. E. Yeast in the March issue of *Metal Progress* was of particular interest because of similar experiences at Pratt & Whitney Aircraft in the magnetic particle inspection of silver-plated and phenolic resin coated parts.

The photographs below show transfers of magnetic particle indications, using continuous and residual methods and with varying thicknesses of silver plate over the defective areas in the steel parts. These pictures show that our findings duplicate those reported in the article, in that 0.0015-in. plate thickness is the maximum at which all discontinuities will be indicated using the wet continuous method, and 0.001-in. thickness the maximum using the wet residual method.

Phenolic resin coatings have presented an additional problem. The coating, which does not exceed 0.001 in. in thickness, acts the same as plating in reducing the effective magnetic flux at a point of leakage, but the smoothness of the coating also causes the indicating medium to form an unstable pattern which will deform and perhaps slide off entirely, even though carefully handled.

Scotch Tape Transfers of Magnetic Particle Indications at Varying Plate Thicknesses. 1800 amp. rectified a.c. for circular magnetization; and 9000 amp.-turn (4-in. diameter opening) solenoid for longitudinal magnetization. Particle suspension 1.5 oz. per gal. in varsol. Line across lower right-hand transfer locates microspecimen showing cracks after grinding 0.003 to 0.013 in.

Residual Method | Continuous Method



This has made it necessary to inspect new parts *before* this coating is applied. In service the surface of the resin film becomes slightly roughened, and no trouble is experienced when inspecting used parts, as far as pattern stability is concerned, or in obtaining good indications of defects in the steel underneath the coating.

The many fractured tools found by Mueller and Yeast are also a familiar problem to us. In our experience, it is always due to residual stresses in the part caused by improper heat treatment or abusive grinding. These residual stresses are released in plating, either by hydrogen embrittlement, or simply through a lapse of time and spontaneous fracture. In either event, tempering or double tempering is helpful in reducing the susceptibility to fracture. A common practice is to temper after grinding but before plating to reduce any residual stresses, and again after plating to remove hydrogen.

Magnetic particle inspection, however, will only show actual cracks. Abusive grinding may produce surface stress conditions which will give rise to cracking when parts are plated. L. P. Tarasov, in his paper "Detection, Causes and Prevention of Injury in Ground Surfaces" (*Transactions* ASM, V. 36, 1946, p. 389) has described techniques of etching which have been very useful in revealing the degree of damage caused by grinding, and thus to study and improve grinding procedures to avoid such damage. By such means, areas liable to crack during, or as result of, plating may be revealed and corrective steps taken.

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Materials Control Laboratory
Pratt & Whitney Aircraft Division
United Aircraft Corp.

Carbide Precipitation in 25-20

Charleston, West Virginia

To the Readers of Metal Progress:

The accompanying photomicrographs were taken of a section from a Type 310 (25-20 Cr-Ni) stainless steel tube which had been in service at 500° C. (930° F.) for approximately 3000 hr. The tube was supplied in the cold drawn and annealed state and was given no heat treatment prior to installation. The specimen was cut from the top of the tube, which had been under no stress during service. It was etched electrolytically in 10% oxalic acid to reveal the carbides.

In this specimen carbide precipitation has occurred along crystallographic directions within the grains as well as at grain boundaries. The change in crystallographic direction at the twin boundaries is clearly shown at lower right in Fig. 2. Individual carbide particles have grown along a particular crystallographic direction in the area shown in Fig. 3. The heavy carbide precipitation at grain boundaries is shown in both.

A specimen from this tube was exposed 50 hr. to a boiling H_2SO_4-$CuSO_4$ solution and examined metallographically to determine whether it was susceptible to intergranular attack. No grain boundary penetration was observed and the specimen withstood a full 180° bend without failure. Apparently, the long period of operation at high temperature had allowed sufficient diffusion to replenish the chromium which had combined with carbon at the grain boundaries. The specimen was thus immune to intergranular corrosion.

David A. Vermilyea
Metallurgist
E. I. du Pont de Nemours & Co.

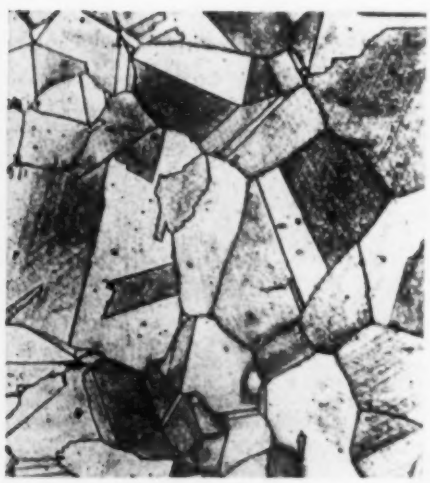


Fig. 1 — Cold Drawn 25-20 Tube After 3000 Hr. at 930° F. Above at 100×, other two at 1500×

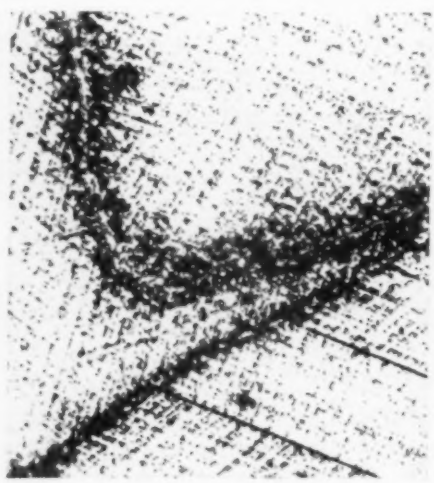


Fig. 2 — Twinned Crystals; Carbides Reveal Changed Crystallographic Directions, Twin to Twin

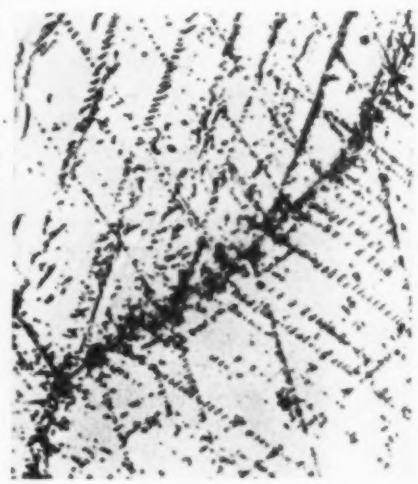


Fig. 3 — Platelets of Carbide Nucleated Along Crystallographic Planes in Adjoining Crystals

British Experimental Blast Furnace

LONDON, ENGLAND

To the Readers of METAL PROGRESS:

A model experimental blast furnace, fully instrumented, and one eighth the size of a normal furnace, has been built at the Stoke, England, works of the Shelton Iron, Steel & Coal Co., Ltd., and is now in operation. It is approximately 14 ft. in height, standing on a base 2½ ft. high for convenience of operation. The internal diameter varies between 2 ft. 2 in. and 2 ft. 7½ in. The furnace was designed by members of Imperial College, London, under the auspices of the British Iron and Steel Research Association and built by the Shelton Co., using routine refractories and a carbon hearth.

Experimental devices incorporated in the design include *(a)* nine thermocouples built in the brickwork — four in the base, two near the hearth, one in the bosh and two in the stack; *(b)* five sampling holes, 1½ in. dia., in either side of the stack, and *(c)* a bosh plate which can be water-cooled if necessary. The water-cooled tuyeres are of steel.

While the furnace is operated along production lines, it is essentially a piece of experimental apparatus and the primary object is research. The measurement of input and output materials with alteration in general variables, therefore, can be carried out as desired.

A view of the furnace is shown on this page. The tap hole and pig beds are in the foreground. The dimensions of the installation may be judged from the operator looking through the tuyere sighting glass, and the 3-ft. rule standing on the bustle pipe and resting against the stack. The platforms on both sides of the furnace immediately above blast-pipe level aid when taking gas samples.

The program of research includes the following:

1. The effect of distribution upon the functioning of the furnace is being determined, with special reference to fuel economy. This involves varying the quantities of charge and order of charging when using either the standard hopper or a new design under development. Small models, scaled to one sixtieth of the size of a normal furnace, indicated the procedure to adopt. The control of stock segregation in the shaft will also be studied.

2. Reactions at various levels in the furnace are being investigated to determine, by correlation with laboratory work, whether a modified temperature gradient is desirable, and the simplest means of exercising such control. For this purpose, the sample holes in the stack are utilized to take gas samples and temperatures simultaneously at various positions across and up and down the furnace. The gas is then immediately analyzed in an adjacent laboratory to determine the percentages of CO_2, CO, H_2 and, if desired, unsaturated hydrocarbons present in the various positions in the stack. These gas analyses will indicate how much material has been reduced at each temperature zone, and it is hoped to assess radiation and cooling losses accurately enough for heat balances to be calculated.

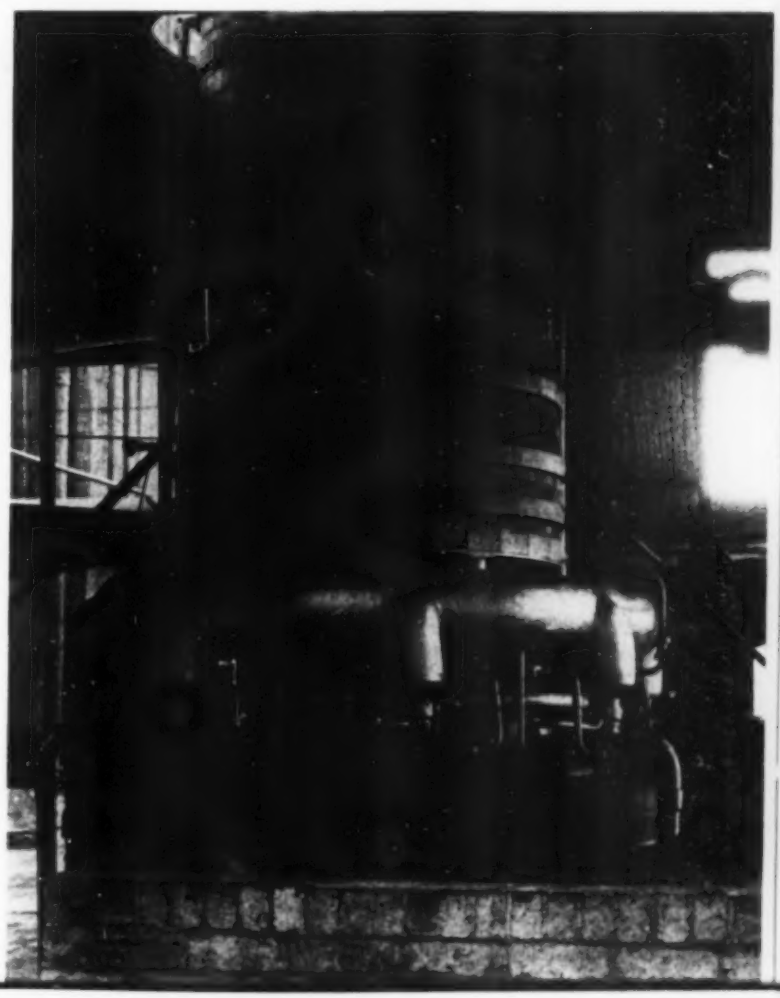
3. The relative reducibility of home and foreign ores under smelting conditions, and the influence of ore grading will be determined. The optimum conditions for sinter utilization will be examined and the influence of coke quality on operation and fuel economy studied.

4. The effect of oxygen in the blast is to be examined. Preliminary study indicates that the main advantages of oxygen will be found in the smelting of ferromanganese and high-silicon irons.

5. Another field for investigation is the control of temperature gradients in the furnace at maximum driving rates, to reduce the depth of the plastic zone of partial fusion, thus facilitating gas flow and minimizing channeling.

The furnace has already had a preliminary trial run of four weeks' duration, during which period encouraging results were obtained, and it was indicated that it was capable of giving results which could be correlated with large-scale practice.

TOM BISHOP
Metallurgical Editor
Iron & Coal Trades Review



The 1948

Open Hearth Conference

ON PAGE 745 of the 20th edition of "Useful Information for Business Men, Mechanics and Engineers" one may learn that oxygen boils at −183.00° C. and that sulphur condenses at 444.6° C.

One of the most effective means of remembering new information is to relate it to what is already known. Much of the program of the 35th Conference of the National Open Hearth Steel Committee of the A.I.M.E., held in Pittsburgh, April 12 to 14, can be related to the simple statement of fact in the first paragraph.

Oxygen and sulphur have many points in common. Both are used in the determination of standard temperatures for calibration of thermocouples; both play a dual role of hero and villain in the efficient production of high quality steel; both occupied a prominent position on the program at the conference.

Sulphur in Steel

Cyril S. Smith has recently revived interest in the old discussion of the relative importance of composition and structure. One key to this problem lies in Longfellow's appraisal, "as the bow is to the bowstring . . . . useless each without the other". There was a time, not long ago, when composition was in the ascendancy. The chemist knew whether or not the sulphur specification was met because he had determined the sulphur content and found it correct. Some old-timers who insisted that added sulphur (or copper) was different from a high residual content were either mildly tolerated or openly regarded as more than a little superstitious.

A paper by L. R. Silliman, on "Effect of Various Resulphurizing Agents on Surface Quality", led to several interesting conclusions: *(a)* Sulphur may be beneficial; *(b)* structure is of considerable importance; *(c)* good practice is that which results in making good steel; *(d)* good steel cannot be defined except by reference to intended use.

The form of the sulphur is important in determining its influence. There is general agreement that in some types of steel low sulphur is desirable. There is much less agreement on what is meant by low sulphur. A shop starting with cupola iron containing 0.200% S treats the metal with soda so that the iron charged into the openhearth contains about 0.100% S. Sometimes the elimination may not be this good — from 0.170 to 0.120%. This shop considers 0.040 or 0.050% in the ingots as low sulphur. It has done an excellent job of sulphur elimination. A mill that starts with 0.030% S, or less, in the iron would consider 0.040% final sulphur to be high. More has been added from the scrap and fuel than has been eliminated by treatment. Both are making quality steel. Apparently total sulphur is an indication of quality for any given conditions of raw material and practice. The fact that a heat has the lowest sulphur content of any made by a given practice may be a better indication of quality than the exact figure of 0.015 or 0.030 or 0.040% S. Each of these values may be high or low depending upon the standard of reference.

Charles W. Briggs read the results of a comprehensive study of sulphur and cited the following effects:

Reported by Frank G. Norris
Statistical Metallurgist
Wheeling Steel Corp.
Steubenville, Ohio

1. Sulphur is an active inclusion-former—the higher the sulphur the greater the number of inclusions.

2. The chain-type inclusion produces low ductility in cast steel, especially as the sulphur content increases above 0.025%.

3. Improper inclusions of the sulphide type make a steel more susceptible to hot tearing.

4. Sulphur decreases the strength and ductility of cast steel at temperatures between 2200 and 2550° F. and increases susceptibility to hot tearing.

5. Increasing sulphur lowers the notched-bar impact value. Type II inclusions produce the greatest effect.

6. Increasing sulphur content increases resistance to grain growth, both with and without aluminum additions.

Most of these effects are related to inclusions or grain size and must be regarded as the effect of structure rather than strictly the effect of composition as such. This distinction offers the possibility of two methods of attack. One is to dodge the problem by maintaining sulphur at such a low level that it can be tolerated. The other approach is to accept the presence of sulphur as unavoidable (with given price level, raw materials and fuel) and to seek methods for a positive control of its influence on structure.

Among the slag and bath conditions that favor low sulphur are:

1. High basicity
2. Low FeO (or high CaO/FeO ratio)
3. High residual manganese
4. Fluid slag, and good action. (High temperature favors both.)
5. High slag volume
6. Low sulphur input, including all materials and fuels.

McKune Award

This year Leo R. Silliman, openhearth metallurgist of the Republic Steel Corp., Buffalo, received the McKune Award (established in 1940 in memory of Frank B. McKune, a charter member of the Open Hearth Committee) for his paper, "The Effect of Open Hearth Practice on Yields and Surface of Rimmed and Semikilled Steel". A systematic analysis of variables was made. Those having no influence were discarded from further study. Important factors were studied singly and in combination. Considering yield of semikilled grades, the most favorable combination of factors is crown tops, no bleeders, and no washing in the pits. Pouring temperature, mold coating and ingot delivery time are important factors related to surface.

Production at Aliquippa

On the first day, the Jones and Laughlin Steel Corp. went all out to demonstrate its pride in some of the most efficient steel-making and steel-finishing facilities in the world by inviting the conference to an all-day plant inspection. The annual capacity of its Aliquippa works is 1.80 million tons of pig iron and 1.76 million tons of steel, and the five openhearth furnaces have established a monthly record of 115,569 net tons. Incidentally, this is no mean figure even for a shop of twice the size. The practice of the Aliquippa works is an object lesson in the extreme versatility of the basic openhearth process. Pig iron capacity greater than the ingot capacity indicates the small amount of scrap consumed. The skillful combination of pig iron and blown metal takes advantage of the large amounts of iron available. Another interesting statistic is that the Aliquippa works has 480 acres. A visitor to this plant leaves with the impression that he has covered most of these acres!

The combination of time and area is not new in expressing openhearth capacity. Tons per hour per square foot of hearth area is sometimes used. Tons per year per acre of plant area is a figure that expresses degree of congestion. Aliquippa has 3675 tons per year per acre.

The Lighter Side

Any discussion will be fuzzy unless there is general agreement on the definition of terms. A formal recognition of the importance of definitions was given in the paper by H. M. Kraner regarding the A.I.S.I. unit roof. There are two answers to the question, "When is a roof not a roof?" One is "When it falls in." The other—"When the patches are bigger than the roof." An openhearth furnace is like the Irishman's shanty because the roof never needs fixing until it falls in and then it is too late. The scheduled rebuild is just a systematic way of fixing the roof in dry weather.

The latest inside information can now be released regarding a new type of radio contest to be sponsored by the General Nuisance Refractory Co. Each week Gen. Nuisance, in person, will select a name at random from the 1947 directory of the Open Hearth Conference. The individual selected will be called by phone and asked the question, "How many of our nonleak stopperheads do you now have in your plant?" One ton of genuine prewar No. 1 heavy melting scrap will be flown into the plant by a B-29 bomber for each stopperhead in the contestant's possession.

When that inimitable master of wit and wisdom, H. L. Tear, takes the floor there is never a dull moment. His discussion of "Evaluation of Factors Affecting Tons Per Melting Hour in Present-Day Operations" established a new high for punch lines per page. The Tear test for heavy melting scrap — "Hold a piece in the air and let go; if it falls, it is heavy melting scrap." "Matching melt carbon with carbon specification helps to improve production rate. The melt carbon is at the mercy of the blast furnace. The carbon ordered is at the mercy of the blooming mill, if that department has any mercy."

Fellowship Dinner

One of the changes for which the 1948 conference may be remembered is that this was the first time a seating list was prepared for the Fellowship Dinner. The speaker, C. E. Wilson, president of General Motors Corp., directed the attention of the group away from technical problems to others more puzzling and of wider interest.

On his topic, "Steel's Not Everything", Mr. Wilson said, "The material progress we have made during the last 100 years was fundamentally based on the discovery of mechanical power and the development of iron and steel. Yet neither of these things could have been accomplished except in a proper political and social environment.

"Now this great fundamental invention, our self-perpetuating free society, which has contributed so much, even in our lifetime, to the prosperity, health and happiness of the American people is being challenged not only by communists, fascists and socialists, but by union monopolies and union dictators as well.

"I am sure I will be accused of being anti-union for what I am saying here this evening, but I am not. The simple fact is that monopoly unions exercise power over the industrial life of the nation. No minority group, not elected by the people and not responsible to the people, can be allowed to exercise such power.

"We can no longer tolerate those subversive elements which are using labor unions to sabotage our production and promote social and political unrest. The progress of our country and the peace of the world depend on how we promote the realism of work as well as the idealism of our free society."

There were, of course, many topics discussed at the conference which are not touched on in this report. The effect of openhearth practice on quality or some measure of performance could be said to be the general topic of the meeting. Scrap preparation and charging, temperature measurement, and oxygen discussions fitted into this general theme. Next year's conference will be held in Chicago, April 18, 19 and 20, 1949.

Crown Cork and Seal Company, Baltimore, Md., uses austenitic chromium-nickel stainless steel to assure the sanitation, durability and trouble-free performance of their CEMAC 28 unit. Easily cleaned, this stainless steel is highly resistant to foodstuffs, atmosphere, most organic and a great many inorganic chemicals, and to dyes and sterilizing solutions.

Use of STAINLESS STEEL.... *minimizes "CLEAN-up" time*

and CUTS MAINTENANCE COSTS, *too!*

You can save money—and time too—because of the ease with which this equipment can be cleaned.

The rotary vacuum filler-capper, shown above, is equipment of this kind. It saves money, time and labor, *day-after-day.*

Easy to clean and keep clean, it also combats wear and general abuse because it's made of chromium-nickel austenitic stainless steel.

Corrosion-resistance combined with high strength make chromium-nickel stainless steel an ideal material for equipment. Considerably stronger than plain carbon steel, it permits cutting bulk and deadweight without sacrificing strength or durability. In addition it may be readily formed, and is easy to weld.

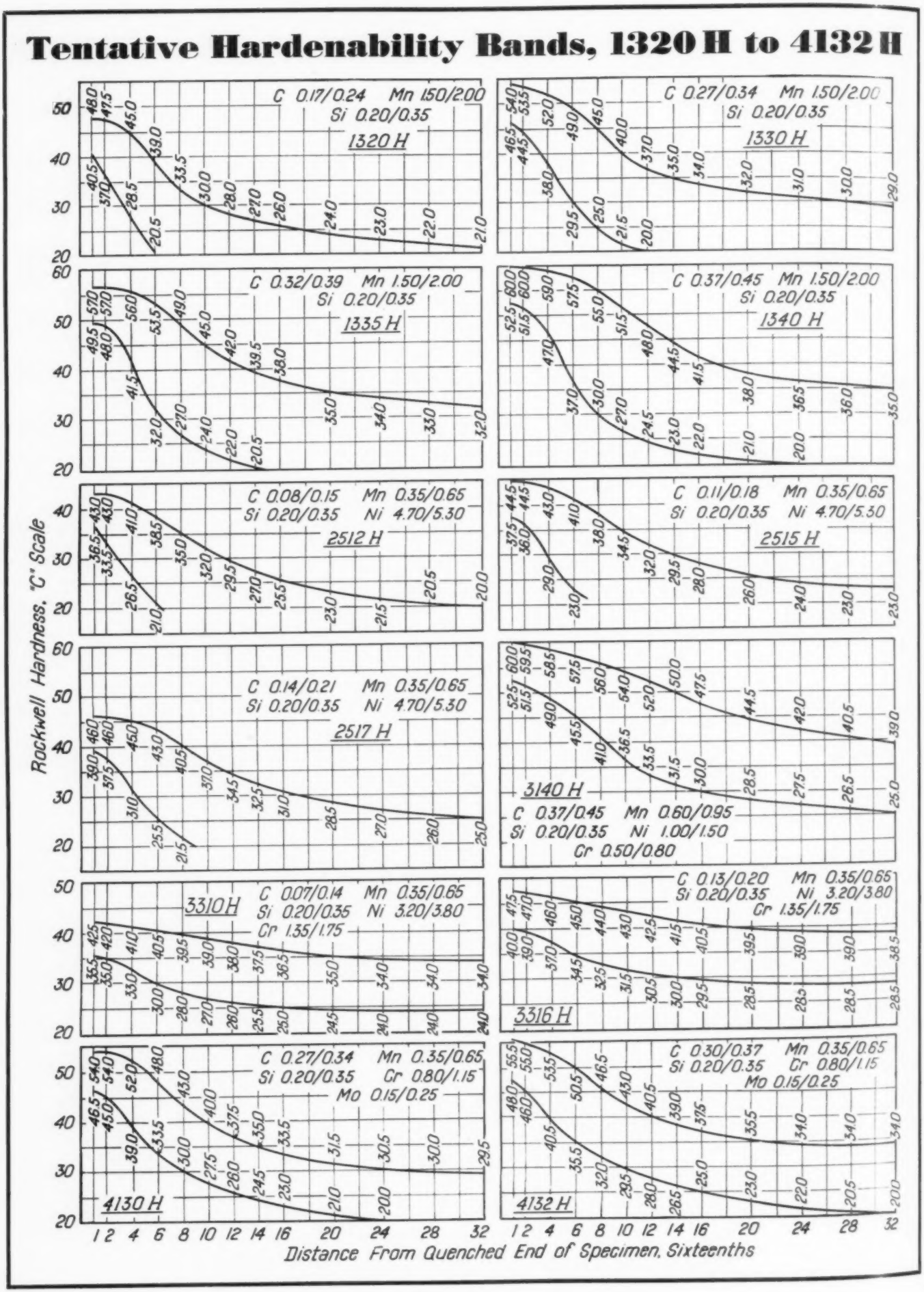
Leading steel companies produce stainless steel containing nickel in tubular, sheet and strip form. A list of the sources of supply will be furnished on request.

EMBLEM OF INCO SERVICE
TRADE MARK

Over the years, International Nickel has accumulated a fund of useful information on the properties, treatment, fabrication and performance of engineering alloy steels, stainless steels, cast irons, brasses and bronzes and other alloys containing nickel. This information is yours for the asking. Write for "List A" of available publications.

THE INTERNATIONAL NICKEL COMPANY, INC. 67 WALL STREET NEW YORK 5, N. Y.

Tentative Hardenability Bands, 1320 H to 4132 H



The establishment of tentative hardenability specifications for alloy steels (a joint project of the Society of Automotive Engineers and the American Iron and Steel Institute) is an event of major importance, and warrants much wider use by metallurgists in the consuming industries.

Hardenability Control for Alloy Steel Parts

THIS DISCUSSION, whose text is the outgrowth of the talks given before many chapters of the ⊜ in several versions during my term as president of the society, pertains to the alloy steel compositions used in this country for the manufacture of machinery. Every machine component must perform certain functions and must possess certain properties to perform these functions properly. The steel used for each component is the one that appears capable of developing those properties most uniformly and at the least cost. No matter what kind of steel is selected, however, it is impossible to use any large quantity of it, involving many heats, without having to contend with considerable differences in behavior during processing and in physical properties. This is because different heats made to the same specification have differing chemical compositions within the limits allowed.

The use of the Jominy hardenability test during the last 10 years has established the connection between variation of chemical composition of the steel and variation in physical properties and in nature of response to heat treatment.

In the first paper* on the end-quench hardenability test, 10 years ago, hardenability of 12 heats of fine-grained 4615 steel was reported as varying from ½ to 2½ in. for Rockwell C-60 (Vickers 700 or Brinell 660). Many doubted that 4615 varied that much; nobody at that time had an explanation. It was not long, however, before Grossmann and others had linked up this hardenability variation with the variation of the chemical composition within the allowable limits; their work† permitted the calculation of hardenability from chemical composition. Now we know that chemical composition on the high side of the specification will cause one set of troubles and chemistry on the low side will cause quite a different set of troubles. Chemistry on the high side will sometimes cause too much or too deep hardening, resulting in cracking, too much distortion and necessity for raising the tempering temperature to bring the final hardness within the desired limits. On the other hand, chemistry on the low side sometimes means insufficient hardening to produce the desired physical properties and will necessitate lowering the tempering temperature.

Just how much variation in hardenability would occur within the chemical limits for each alloy steel specification was the problem during the war set before a committee sponsored jointly by the Society of Automotive Engineers and the American Iron and Steel Institute. Hardenability

*"A Hardenability Test for Carburizing Steel", by W. E. Jominy and A. L. Boegehold, *Transactions* ⊜, V. 26, 1938, p. 574.

†Summarized in "Metal Progress Data Sheets" (1946 Edition) No. 51, 52, 55 and 57.

By A. L. Boegehold
Past President ⊜
Head, Metallurgy Department
Research Laboratories Division, General Motors Corp.
Detroit

curves on hundreds of heats of each steel were collected and studied. The spread of hardenability calculated from chemistry indicated that the variation for any current specification would be great, but the spread shown by actual hardenability tests was even wider, due to the residual alloying elements in some heats. It seemed obvious that steps should be taken to protect the steel user by restricting hardenability within narrower limits, and tentative hardenability limits were set up. With some changes these are now available for 62 different alloy steels. These tentative standards have been published in June 1947 by the S.A.E. and A.I.S.I. jointly in "Contributions to the Metallurgy of Steel — No. 11; Hardenability of Alloy Steels".*

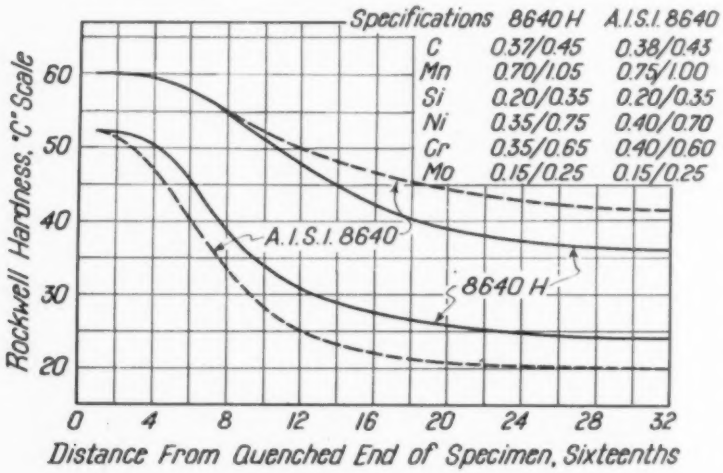


Fig. 1 — Typical Spread of Representative End-Quench Hardenability Curves for Steels Ordered to Chemistry and to Hardenability Specifications

The difference between the new hardenability limits in an "H" steel, so-called, and the spread of hardenability incurred by specifying according to chemistry by A.I.S.I. or S.A.E. standard specification, is illustrated in Fig. 1. Two dotted hardenability curves show the spread of hardenability encountered in a large number of heats of 8640 ordered to chemistry limits, and two solid hardenability curves show the limits of hardenability guaranteed by the steelmaker if 8640H is ordered.

The method and statistics on several hundred heats of each type of steel used in deciding on the location of the limits for the new H-bands have been described by John Mitchell in his "Progress Report on Hardenability Bands" in the 1945 Year Book of the American Iron and Steel Institute.

Over a period of many years, steelmakers found that about 7% of the heats made failed to meet the chemistry specifications and had to be diverted to other uses. In the study of the hardenability steels it was therefore decided to discard 7% of the heats instead of widening out the hardenability band on the high or low side to include them. However, the heats that fall outside the H-band limits and have to be diverted are not the same heats that fall outside the chemistry specifications. Since the number of rejected or diverted heats involved is no greater than formerly, the customer can specify the narrower H-band limits without paying any extra. Notice also that the chemistry limits allowed for the H-band specification — a typical example is noted in Fig. 1 — are wider than the limits when ordering to chemistry alone. This does not mean that *all* elements could be at the low limits or high limits in one heat of steel; that would result in hardenability outside the band limits. It means that chemistry can be anywhere within the limits shown, if *at the same time* the hardenability curve is inside the H-band limits. The way it works out is that most of the chemical elements are toward the middle of the range, while one or two of them may be near the high or low limit for that element. Steel ordered to chemistry limits could have all elements but one toward the middle of the range and one outside the limit, and therefore would be rejectable, yet the hardenability of such a heat would be well within the H-band range and would be a perfectly satisfactory steel for the purpose intended. As another example, the chemistry specification would permit all the chemical elements in one heat to be on the low limit, or on the high limit, thus presenting a problem involving too low or too high hardenability. This cannot happen when steel is ordered by the H specification.

On this basis there is really no justification for ordering steel to the old chemistry limits.

The H-band limits have been available to consumers for nearly two years. Many metallurgists have made no attempt to use them. Of those who have used them, some affirm that the closer hardenability limits result in more uniform material and less trouble in the shop, others (who have to do precision heat treating) report that the H-band limits are still too wide, others say that the old chemistry specification plus a minimum hardenability specification gives them all the control needed, while still others say that the old chemistry limits are just as good. If surface hardness must be under accurate control, the narrower carbon range obtained with the chem-

*These revised H-bands will be published as data sheets in *Metal Progress;* the first is in this issue, page 696-B.

istry specification is frequently preferred. The first and second statements are justified but the third, fourth and fifth are not. It is true that certain parts must be held within narrower limits than can be supplied by the tentative H-band specification, but it is also true that where satisfactory steel can be ordered to standard chemistry limits a substitution of the H-band specification will give better control and less trouble in the long run.

The carbon range of the chemistry specification for the standard openhearth and electric furnace alloy steels (A.I.S.I. and S.A.E.) is five points. Applied to the range in more than one heat this could actually be nine points when two points allowance for check analysis above and below the limits is included. The total range for the H-band specification on this basis is 12 points. The fact is, however, that the steelmaker cannot allow that much variation in carbon, for if he did the hardenability curve at the $\frac{1}{16}$-in. location would not be within the prescribed limits. Note that, in Fig. 1, the maximum and minimum spread for 8640H is no wider than when the chemistry specification is used. As a matter of fact, to illustrate with 8640, the steel whose H-band is shown in Fig. 1, the actual hardness range corresponding to the carbon range permitted by the chemical specification 0.36 to 0.45 would be C-50 to 62. We see, therefore, that ordering steel to chemistry will result in *more* hardness spread at the surface of hardened articles than would occur when the steel is ordered to the hardenability limits.

As to the use of a chemistry specification *plus* a minimum hardenability, it should be noted that no advantage is gained that cannot be obtained with the H-band specification. The proposed alternative method sacrifices control on the high side of the band, and unduly high hardenability often gives cracking and distortion trouble.

The principal misconceptions regarding H-bands have arisen because conclusions are drawn from its use within a limited set of conditions. For example, if some part of small section size is made from a steel having more hardenability than needed for that size section, then there will seem to be no difference whatsoever between results with steel bought to chemistry limits and steel bought to H-band limits. Especially will this be true if hardness tests are made only at the surface of that part, and when results from too few heats are considered. But a broader view must be taken. Later on, two charts will be discussed showing how the limited aspect just mentioned is related to the whole problem.

In order to get a better perspective of the over-all problem, it should be studied by curves showing hardness across the cross section (hardness traverses) as an adjunct to the end-quench hardenability curves. The time and expense connected with actually determining hardness across

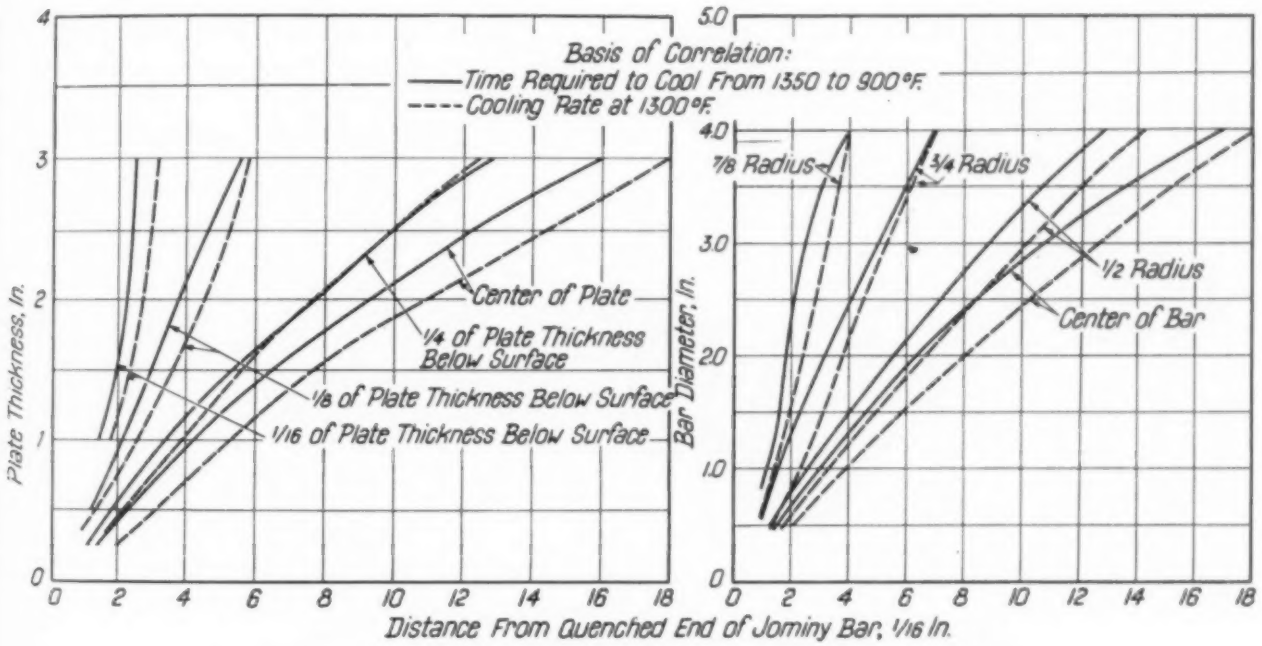

Basis of Correlation:
Time Required to Cool From 1350 to 900°F.
Cooling Rate at 1300°F.
Plate Thickness, In.
Center of Plate
1/4 of Plate Thickness Below Surface
1/8 of Plate Thickness Below Surface
1/16 of Plate Thickness Below Surface
1/8 Radius
3/4 Radius
1/2 Radius
Center of Bar
Bar Diameter, In.
Distance From Quenched End of Jominy Bar, 1/16 In.

Fig. 2 — Correlation of Cooling Rates and Times Between Jominy End-Quench Test Pieces and Water Quenched Plates (Left) and Bars of Various Diameters

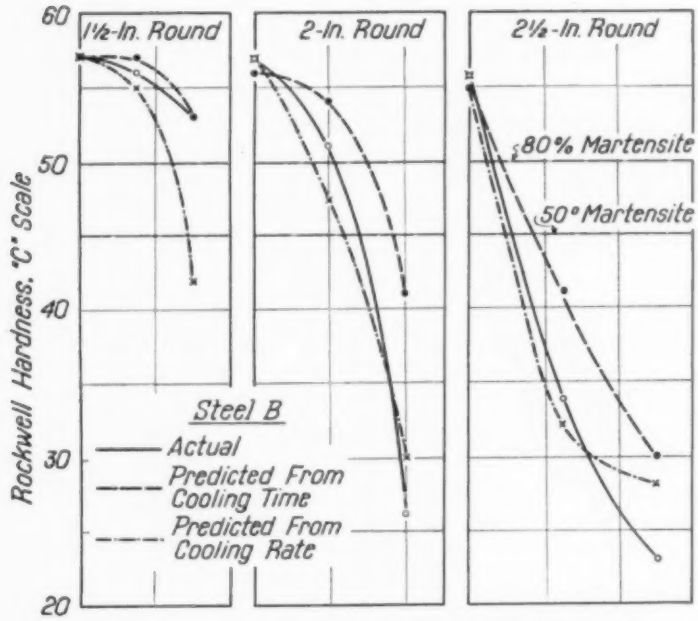


Fig. 3 — Actual Hardness on Cross Sections of Rounds Compared With Hardness Predicted by Cooling-Time and Cooling-Rate Correlations. Manning's Steel B; composition: 0.44% C, 0.91% Mn, 0.22% Si, 0.09% Ni, 0.01% Cr, 0.05% Mo

rates at 1300° F. These curves were published in *Metal Progress* Data Sheet for January 1947, and are used for deriving hardness versus cross section curves.

Sometimes we hear doubts expressed about the accuracy of such hardness-depth curves when derived from the Jominy curve. We should, therefore, digress for a few moments to establish its validity. Likewise some workers on hardenability have questioned the practice of using the cooling rate at 1300° F. as an indication of the effectiveness of a given cooling cycle, contending that the cooling speed below 1300° F. may affect the final hardness.

Wilks, Cook and Avery, in their paper entitled "Further Developments of the End-Quench Hardenability Test" in *Transactions*, V. 35, 1945, p. 1, believe that the entire cooling curve must be considered. With this in mind, they developed a new method for testing hardenability. They used a bar quenched on both ends and varied the length of the bar to represent the diameter or thickness of the section under investigation. Their method does duplicate cooling curves in the quenched section and in the test bar, but it is at a disadvantage because it does not provide a *standard* test bar but uses a different length of test piece for every diameter of bar that is in production.

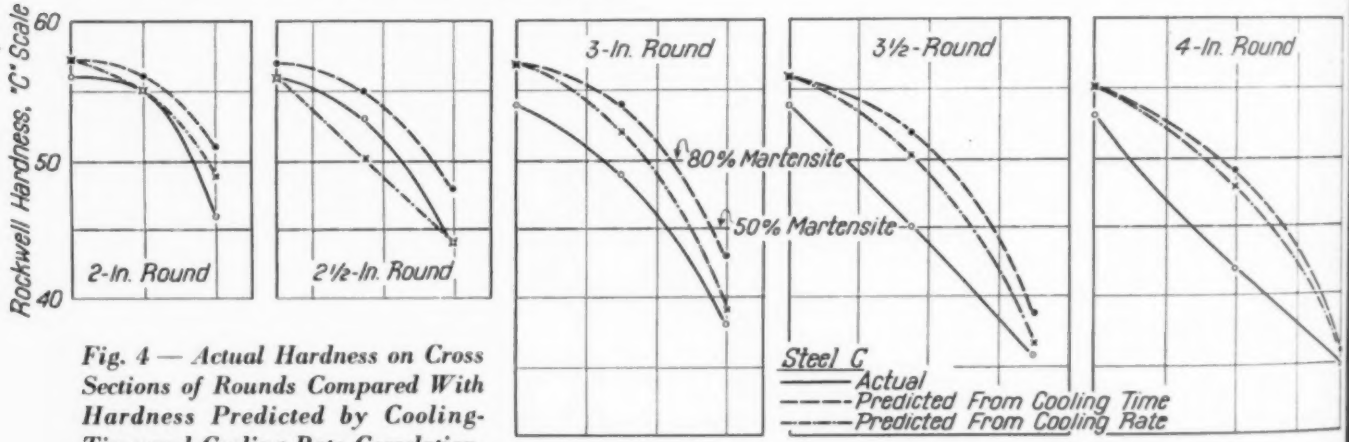


Fig. 4 — Actual Hardness on Cross Sections of Rounds Compared With Hardness Predicted by Cooling-Time and Cooling-Rate Correlation. Manning's Steel C; composition: 0.43% C, 0.96% Mn, 0.25% Si, 0.57% Ni, 0.54% Cr, 0.27% Mo

the section can be saved by deriving these curves from the Jominy hardenability curves. (Often only the hardenability curve and not the steel itself is available to the metallurgist who must decide whether it will be suitable for some part.)

Derived hardness curves across the section are obtained by assuming that any point within a quenched section which has the same cooling rate as some point on the Jominy bar will also have the same hardness as that point on the Jominy bar after quenching. Curves are available correlating points on the Jominy bar with points in quenched sections of various sizes having the same cooling

It is admitted that the speed of cooling after passing below 1300° F. may determine final hardness, but studies of the effect of cooling below 1300° F. should be made on the present standard Jominy hardenability bar so that data derived from hardness curves and cooling rates can be compared. Therefore, a new correlation between Jominy bar and quenched sections was prepared in which time to cool from 1350 to 900° F. was used instead of the cooling rate at 1300° F. This pertains to water quenched rounds and plates,

shown in Fig. 2, comparing the data from experiments based on the two assumptions. Both sets of curves were based on newly determined cooling curves in hardenability bars and in quenched sections.*

All test bars in these tests, both end quenched and submerged quenched, were made from the same composition steel (N.E.9445) and cooling curves were taken with highly sensitive instruments. It is evident from Fig. 2 that higher hardness for a given size section will be predicted from the cooling-time correlation (that is, the time required to cool from 1350 to 900° F.). Which method should be used? The accuracy of predicting hardness from the cooling-time correlation was carefully checked by George K. Manning in *Metal Progress* for October 1946, in rounds from 1 to 4 in. diameter made from four different steels. He did not include hardness-depth curves derived from the new cooling-rate correlation, so I have prepared these curves for three of his steels from the relations shown in Fig. 2, and plotted them in Fig. 3, 4 and 5, together with curves derived from the cooling-time correlation.

In the 1-in. and 1.5-in. round, the hardness predicted from the cooling-time correlation was very close to the actual hardness. In the sizes from 2 to 4 in., however, the hardnesses predicted from the cooling-rate correlation are closer to actual hardnesses than those predicted from the cooling-time correlation.

The biggest discrepancy was 15 points on the Rockwell C-scale at the center of the 2-in. round of Steel B, Fig. 3. Manning's hardenability curve for Steel B showed a very steep gradient in that part of the Jominy bar corresponding to the center of a 2-in. round quenched in water. This, of course, increases the possibility of error; it can be stated that, as a general rule, predictions made from the steep part of the hardenability curve will tend to be more in error than predictions from the flatter parts of the curve.

Figures 4 and 5 show curves for sizes up to 4 in. in diameter. Our own work at General Motors' Research Laboratory Division has also shown that the cooling-time correlation is more accurate in sizes up to 1½ in. and that in heavier sections the cooling-rate correlation will predict more accurately.

The reason for this is not yet apparent. Manning suggests that stresses (due to thermal gradients and to sequence of volume changes accompanying transformations) may alter the amount of transformation and thus affect the final hardness of the mixed microstructure.

No doubt some of the discrepancies can be caused in this way.

Although these derived hardness-depth curves are not 100% accurate, they are adequate for practical purposes because the accuracy is sufficient to judge the suitability of a steel for a given usage. The method described above, whatever its shortcomings, is the most accurate method available to the metallurgist for judging the application of a steel to the machine part to be made. Compared to the practice, followed only 10 years ago, of using one set of hardness versus cross section curves made on one heat of steel to represent all heats of the same S.A.E. specification, the present method of predicting is highly accurate and infinitely less likely to lead us into erroneous results. We can say from considerable experience with this method during the war that the results are highly satisfactory.

*The entire data were presented in Office of Scientific Research and Development's O.S.R.D. Report No. 4386, Serial M-404, entitled "Heat Treatment of National Emergency Steels for Use in Tanks, Combat Cars, Gun Mounts and Other Ordnance Materiel". Fig. 2 on p. 699 is from *Metal Progress* Data Sheet, January 1947, p. 96-B (based on data therefrom).

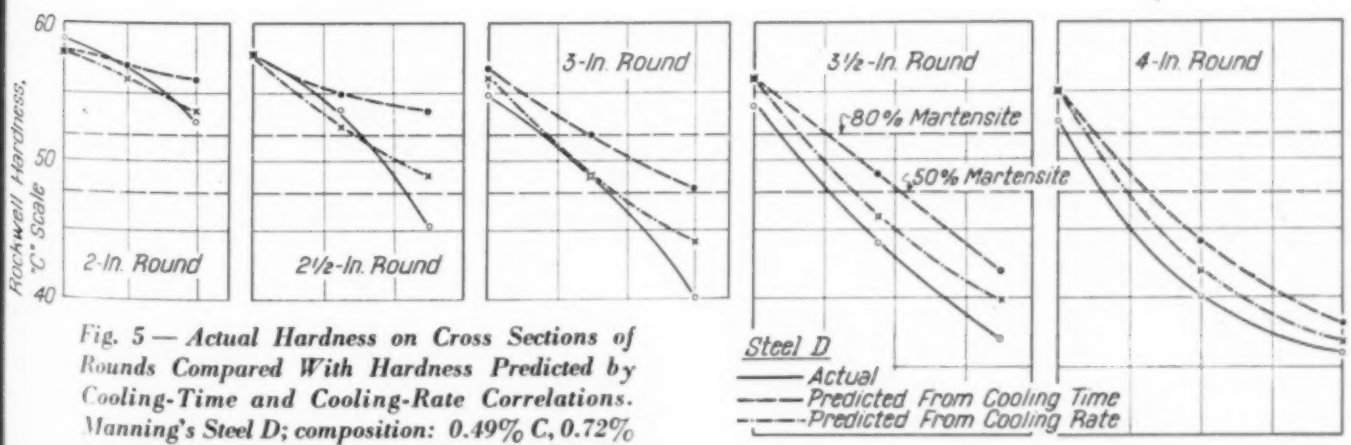


Fig. 5 — Actual Hardness on Cross Sections of Rounds Compared With Hardness Predicted by Cooling-Time and Cooling-Rate Correlations. Manning's Steel D; composition: 0.49% C, 0.72% Mn, 0.30% Si, 0.52% Ni, 0.54% Cr, 0.20% Mo

Severity of Quench

At one time we believed it would be necessary to have separate correlations between quenched sections and the Jominy bar to represent different severities of quench as a result of varying the velocity of the quenching medium from 0 to 1000 ft. per min.

The correlation between the hardenability bar and water quenched rounds, shown at the right side of Fig. 2, is based on water flowing past the surface of the piece being quenched at 200 ft. per min. (3⅓ ft. per sec.). We found that speeding up the quenching stream has much less effect on speed of cooling than is commonly thought. The cooling rate is not changed appreciably at the center of the bar or plate, and is only slightly affected at half-way between the surface and center for either plates or rounds. Greater speed of cooling occurs between the surface and half-radius, and the increase in hardness that can be expected by increasing the velocity of the quenching medium is shown in Fig. 6, above.

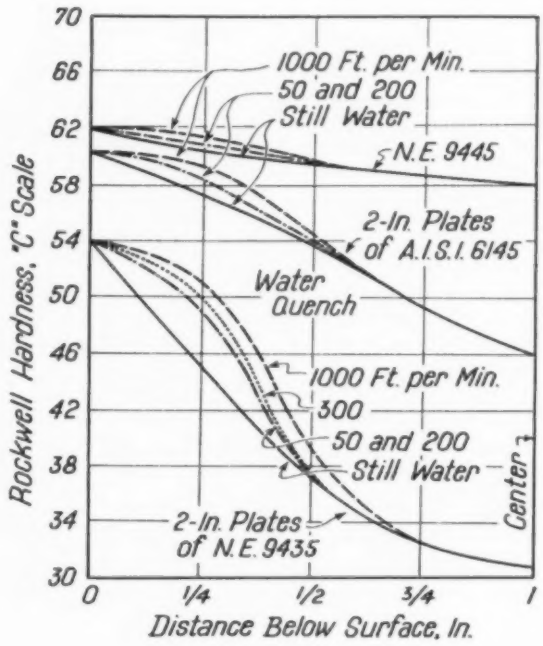


Fig. 6 — Increasing the Velocity of the Quenching Stream Increases the Hardness of the Outer Layers Only and to a Noticeable Degree Only in Steels of Low Hardenability

Four hardness curves between surface and center of a 2-in. plate are predicted for N.E.9435 steel, one for still water and one each for 50, 300 and 1000 ft. per min. The cooling rates at 50 and 200-ft. per min. water velocity are the same. Three hardness curves are predicted for A.I.S.I. 6145 and for N.E.9445 steel. The influence of velocity of quenching water depends upon the degree of hardenability possessed by the steel. In the case of N.E.9445, which has high hardenability, the increased water velocity has very little effect because even when quenching in still water, the cooling rates in a 2-in. plate are enough to produce almost the maximum hardness obtainable.

On the other hand, the steep part of the hardenability curve of relatively shallow hardening steel, like N.E.9435, is at that part of the Jominy bar having cooling rates the same as those midway between surface and center of a 2-in. plate, water quenched. Therefore, an increase of cooling rate by speeding up the quench water produces a relatively large hardness increase in this region but increases the depth of hardening only $\frac{3}{16}$ in. The increased water velocity in no way compares with the effect of increased alloy and carbon in producing through-hardening. A steel like 1045, having the steep part of its hardenability curve at cooling speeds comparable to those near the surface of the 2-in. plate, would show a big increase in hardness at the surface of a 2-in. plate when the quench water speed is increased greatly — to 1000 ft. per min.

The conclusion we reached from these data was that it would not be worthwhile to go to all the trouble and work necessary to establish separate correlations between cooling rates in the Jominy bar and in quenched sections for various water velocities because the hardness predicted from the present correlation is sufficiently close. It is enough to know that only slightly greater penetration can be obtained by drastically increasing the quenchant velocity.

The most important benefit from high velocity in the quenching medium is insurance of maximum hardness and freedom from soft spots at the surface and just below the surface. If the steel is one with a hardenability curve having a steep gradient in that region of the Jominy bar having cooling rates like those near the surface of a 2-in. plate, a speed-up of the quenching water velocity will produce a marked increase in surface hardness of a quenched 2-in. plate. It is only under these critical circumstances that hardness is largely increased by speeding up the circulation of the quenching water.

From the foregoing discussion of derived curves for hardness across the section we can, with a degree of accuracy shown in the comparison with Manning's result, use such derivations to predict the usefulness of a steel in different section sizes and to illustrate the advantages of ordering steel to the H-band specification. Figure 7 shows a number of such hardness curves (frequently called "U-curves" from their complete shape) derived from the maximum and minimum hardenability curves bounding the hardenability band for 8745H steel. These curves show exactly what can

be expected much more understandably than does the hardenability band by itself.

Hardness Across the Section

In Fig. 7 the dotted horizontal lines give the percentages of martensite present in the as-quenched structure.* It is obvious that, in the sections larger than 1½-in. round oil quenched, and 2-in. round water quenched, even with the more restricted range of hardenability obtained with the H-band, we have to contend with greatly different structures and hardness in steels representing the high side and the low side of the hardenability range. For example, a 2-in. round, oil quenched, will have 95% martensite and C-56 at the center if made from steel on the high limit of hardenability, but, when made from steel on the low limit, the center hardness would be only C-40 and contain only 20% martensite. Such differences in structural transformations would undoubtedly cause greatly different stress patterns after hardening.

The location of the low limit hardenability curve, of course, determines the sizes to which the steel must be limited, depending upon the requirements of the piece to be made. Stated in another way, the steel selected for a certain part must have a hardenability range whose minimum will produce the structure and hardness required in the part. The minimum hardenability of a steel obviously determines how large a section it can be used for. (These statements apply only where many heats of the same steel are used for making the same piece in production.)

There is no one minimum hardness or structure that must be obtained at the center of a section after quenching; different parts made from alloy steel involve different requirements for as-quenched hardness and structure. For example, a bolt is subjected principally to tensile stress more or less uniformly across its section, and so it is desirable to have uniform structure and strength all across its section; consequently we usually require an as-quenched minimum hardness of C-52, equivalent to 90% martensite. The maximum size bar that could be made from a given steel and still show C-52 at the center, after quenching, is the critical size for that particular steel and is determined from

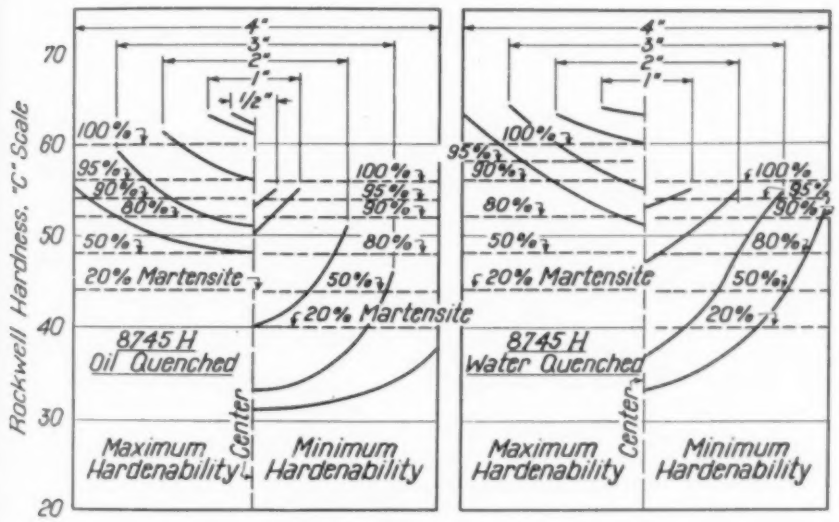


Fig. 7 — Derived U-Curves (Hardness Traverses) for Various Sized Rounds of 8745 H Steel Having Maximum and Minimum Allowable Hardenability

*Proportion of microstructure can be estimated by methods developed by J. M. Hodge and M. A. Orehoski and described in papers on "Hardenability Effect in Relation to the Percentage of Martensite" presented before the American Institute of Mining and Metallurgical Engineers (Technical Publications No. 1800 and 1994).

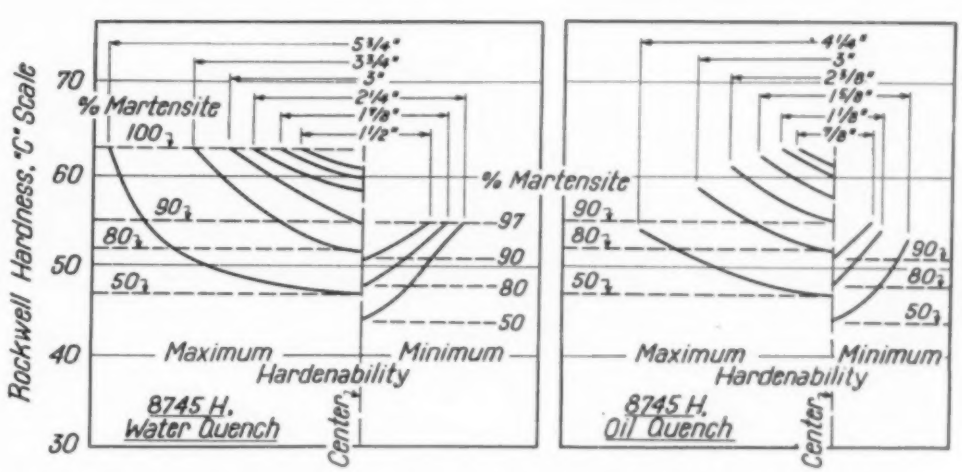


Fig. 8 — Traverses of 8745 H Rounds of Such Size as to Quench Out to 50, 80 and 90% Martensite at Center

the minimum hardenability of the steel. For 8745H (Fig. 7) 90% martensite at the center when oil quenched results in a critical size of about ⅞-in. round.

A method for plotting cross-section hardness curves to show how the critical size changes, depending on what hardness or structure is specified at the center, as-quenched, is shown in Fig. 8. The hardness-depth curves are for sizes which require at least 50, 80 and 90% martensite at the center. This requirement depends on the views of the metallurgist and the nature of the part to be hardened. A rather extensive tabulation of maximum size of bars of the H steels which can quench to specified center hardness, and the corresponding amount of martensite in the structure, was carried as a data sheet in *Metal Progress* for February 1947.

The minimum requirement of 90% martensite in the as-quenched microstructure has been mentioned as applicable to bolts or other tension members. Martensite 80% minimum, equivalent of C-48 for carbon on the low side of 8745H steel, is typical of at least two manufacturers' requirements for automobile axle shafts after hardening and before tempering. Although an axle shaft stressed in torsion and in bending is subject theoretically to no stress at the center, it is necessary to set a fairly high minimum central hardness to be sure that the surface hardness, as-quenched, will be high enough to yield the best structure after tempering to resist the high stresses there.

A minimum of 50% martensite at the center (C-44) is considered by most metallurgists to be inadequate for through-hardened axle shafts. Most auto engine crankshafts and other parts not so highly stressed, or machine parts where the final hardness after tempering is in the lower brackets, will be satisfactory with 50% martensite or less, as-quenched.

The critical sizes for oil quenched 8745H steel for these three structural requirements then becomes ⅞ in. for 90% martensitic structure, 1⅛ in. diameter for 80% and 1⅝ in. for 50%. These maximum sizes are based on structures obtainable in a heat of 8745H having low limit hardenability.

All heats of 8745H specification steel should be limited to use in the critical size peculiar to minimum hardenability, in spite of the fact that most of the heats supplied would give the required structure in much bigger sizes (up to those shown at the left in the charts of Fig. 8). In the 1⅛-in. round, representing use as an axle shaft, the total as-quenched hardness variation from low to high limit hardenability is 15 points on the Rockwell "C" scale from surface of bar with high-side hardenability to center of bar with low-side hardenability. The spread of surface hardness is only 9 points. This is the reason that tempering temperatures must be shifted frequently to make the tempered hardness come within specified limits.

It would be a great boon to steel treaters and metallurgists, especially those using continuous

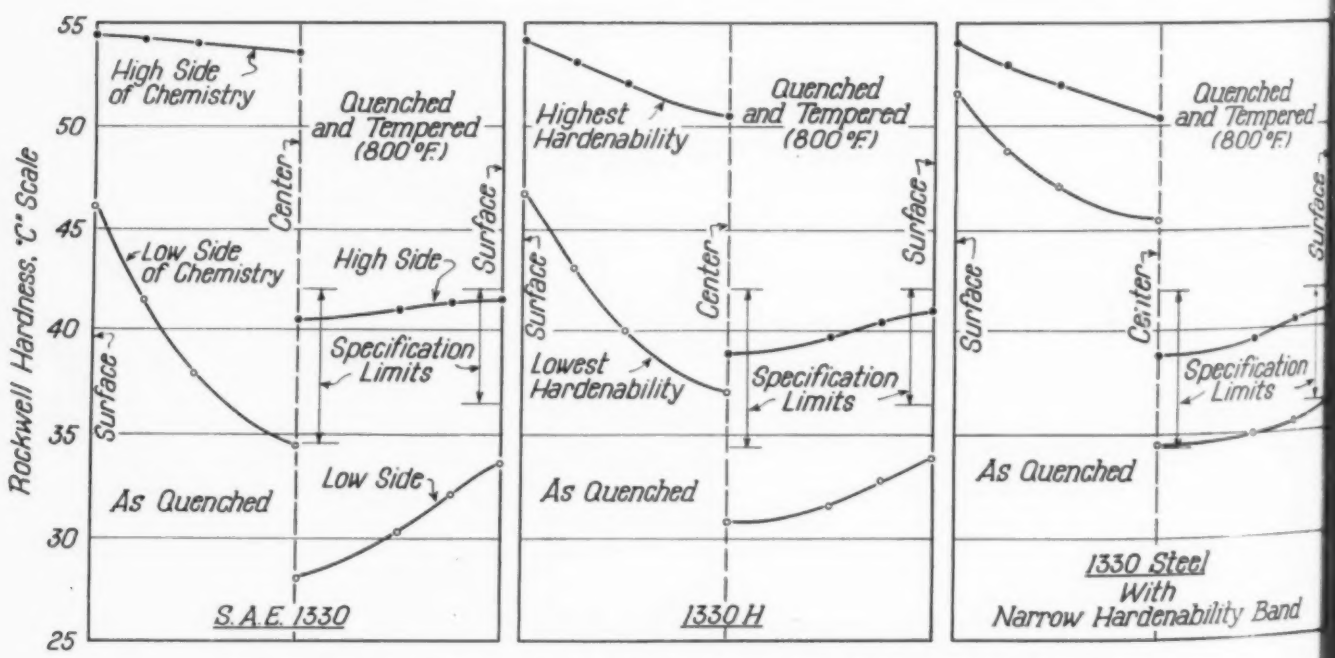


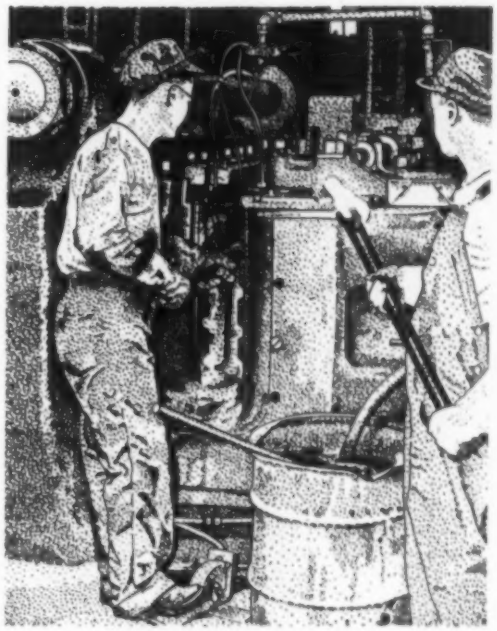
Fig. 9 — Hardness Traverses of Auto Axles Made With Steel on High and Low Side of S.A.E. 1330, 1330H, and of Selected Heats With Narrow Hardenability Band

conveyer furnaces, if they were not required to change tempering temperatures to compensate for varying as-quenched hardness. The width of the hardenability band and the hardness limits specified for the heat treated article determine the adjustments required for tempering temperatures.

Steel for Axle Shaft

Consider this question in connection with through-hardened and tempered axle shafts. In a practice recently used, axles were made of S.A.E.1330, water quenched and tempered to C-36.5 to 42.0 at the surface and C-34.5 to 42.0 at the center. The shafts varied in size, but to illustrate the problem, the hardness relationships are presented in hardness traverses showing the hardness range that would be experienced for heats of water quenched S.A.E.1330 bought to chemistry only, and hardness resulting from tempering at 800° F. Obviously, the as-quenched hardness range is too wide, since the center hardness of steel on the low limit of hardness is at the minimum required after tempering. This means that *no* tempering could be done on such a shaft.

Similarly in the middle chart is shown the as-quenched hardness range for 1330H, and the hardness after tempering at 800° F. again outside the prescribed hardness limits. A lower tempering temperature for low-hardenability heats would suffice to bring all tempered hardness within limits. This is done regularly to compensate for wide variations in as-quenched hardness. This involves checking every heat of steel to determine the proper tempering temperature and keeping heats separate all through processing. In one plant where axles were made from S.A.E.1330 steel, bought to chemistry limits, hardness limits were maintained only at the surface, and to maintain the C-36.5 to 42 limit, it was necessary to vary tempering temperatures from 775 to 875° F. Hardenability band limits would narrow this a little but would not eliminate the necessity of changing tempering temperatures. What the hardenability band specification *would* accomplish would be to avoid heats with such low hardenability that they could not be used for axles, as sometimes occurs when steel is bought to chemistry limits only.



By limiting the range of hardenability to that represented in the right-hand chart it would be possible to temper all heats at the same temperature and meet the final hardness limits. Observe that all hardness values after tempering come within the prescribed limits of 34.5 min. at center to 42 max. at surface. A hardenability band to produce such results would have a spread of 3 points on the Rockwell "C" scale at the $\frac{1}{16}$-in. position on the Jominy end-quench test bar and 8 points C at $\frac{5}{16}$ in. This represents just about the range that will include 66% of the heats made to a chemistry specification. If hardness were tested only at the surface, a slightly wider hardenability band would suffice.

Specification of a hardenability range of this amount is not a possibility with the present state of development of the art of hardenability

Table I — Comparison of Hardnesses to Be Expected in Quenched Rounds of Typical Steels Specified by Chemical Analysis and by Hardenability Bands

Bar Dia.	Hardness at Surface				Hardness at Center			
	Maximum		Minimum		Maximum		Minimum	
	Chem.	"H"	Chem.	"H"	Chem.	"H"	Chem.	"H"
	4340 Steel* (1.75 Ni, 0.80 Cr, 0.25 Mo)							
½ in.	C-60	C-60	C-54	C-52.5	C-59.5	C-60	C-52	C-52.5
¾	59.5	60	53	52.5	59	60	51	52.5
1	59.5	60	53	52.5	59	60	50	52.5
1¼	59.5	60	52.5	52.5	58.5	60	48.5	52.0
1½	59.5	60	52.0	52.5	58.5	60	47.5	52.0
2	59.0	60	51.0	52.5	58	60	45	51.5
3	58.5	60	48.0	52.0	58	60	40	49
	8640 Steel (0.90 Mn, 0.55 Ni, 0.50 Cr, 0.20 Mo)							
½ in.	C-60	C-60	C-51	C-51.5	C-60	C-60	C-49	C-50
¾	60	60	50.5	51.5	59	58	43	46
1	59.5	59.5	49	50	58	57	41	44
1¼	59.5	59.5	49	50	55.5	53	35	39
1½	59	59	48	49.5	54.5	52	32.5	37
2	59	58.5	45	47.5	51	47.5	26	32

*Results of 159 heats bought to chemical specification.

control, unless circumstances are such that heats can be selected.

Table I shows a further comparison between hardnesses in as-quenched sections of heats of two alloy steels bought to chemistry and those bought to hardenability specifications. These figures indicate that if a surface hardness of C-50 is required on a 1¼-in. diameter section, it can be obtained with the 8640H specification, but S.A.E.8640 ordered to chemistry would only apply up to a ⅞-in. round at this hardness — it would be necessary to use a steel with more alloy or more carbon, and either one would cost more money either for the steel or the machining.

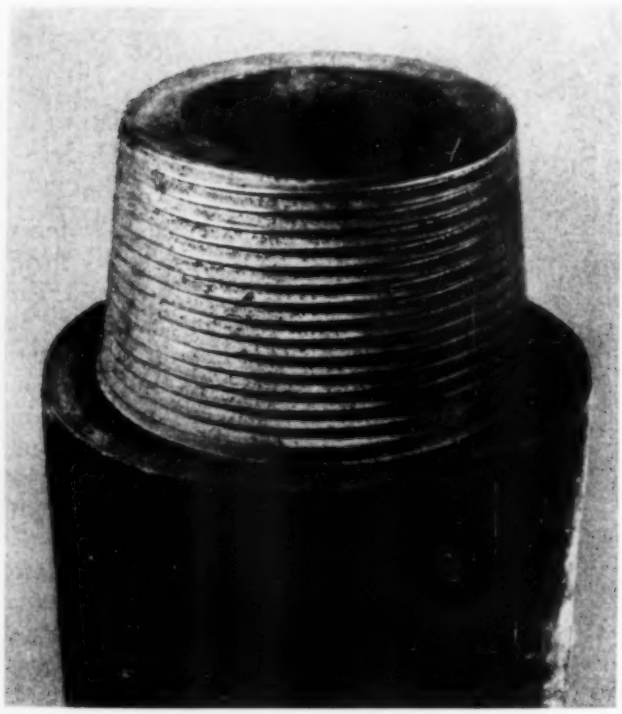


Fig. 10 — Tool Joint Susceptible to Quench Cracks Unless Steel Is Selected to Unusually Narrow Hardenability Band

Suppose we turn to the 4340 steel in Table I. If, in a ½-in. diameter section, oil quenched, a hardness of C-50 is required at the center, either a hardenability or chemistry specification for 4340 would meet the requirement. However, C-50 at the center of a ½-in. section can be obtained with 8640 bought to the H-band specification; to specify the higher alloy and more expensive S.A.E.4340 for this requirement is uneconomical. S.A.E.8640 bought to chemistry, however, would show only C-48 for a minimum hardenability heat and would, therefore, not meet the requirement. This is an example of the rule that the least expensive means of obtaining desired properties is by means of a hardenability specification.

As another example, if C-52 were wanted at surface of a 3-in. diameter, 4340H would do it, but S.A.E.4340 would produce that hardness only in a 1½-in. diameter with a low-side chemistry heat. Again, if C-50 at center of a 2¾-in. section were wanted, 4340H would do, but S.A.E.4340 (bought to chemistry) would allow heats that could reach this hardness only in 1-in. rounds.

A little study along these lines makes it obvious that it will cost the consumer money to buy steel on chemistry.

Examples Where H-Bands Do Not Apply

A few instances will be discussed to show there are circumstances wherein the new hardenability band specifications are inadequate for meeting the requirements.

First let us consider the use of hardenability limits for steels for which no H-bands exist, specifically the selection of steel for a torsion bar spring for heavy vehicles. This part has been successfully made from A.I.S.I.9262 plus 0.25% molybdenum, or from 9262 intensified with Grainal. Hardness specified is C-47 to 51 after tempering at a minimum of 850° F. (This temperature insures adequate ductility for a presetting operation.) Therefore, the minimum as-quenched hardness must be at least C-55. Since the maximum obtainable hardness for 9262 (2% Si, 0.32% Cr) is not more than eight points higher, it is necessary only to specify a minimum hardenability to insure at least C-55 at the center of the torsion bar. One size used was $2\frac{3}{16}$ in. diameter, and from a set of curves similar to Fig. 2 yet which apply to oil quenching, it is found that the center cooling rate is the same as that at a point 14/16 in. from the end of the Jominy bar. The minimum hardenability, therefore, will be represented by a hardenability band whose lower boundary passes through C-55 at 14/16 in. That is all the specification that is necessary (plus the chemistry).

This kind of specification is permissible when there has been no hardenability band established for the steel used, which is true for A.I.S.I.9262 plus 0.25% molybdenum or for A.I.S.I.9262 intensified with Grainal.

Earlier in this discussion, reference was made to the danger of cracking with steel on the high side of the hardenability range. In this particular example there was danger of cracked torsion bars made of steel with any amount of hardenability down to the minimum required. In production, cracking was prevented by a time quench and a controlled time-temperature cycle prior to temper-

ing. Cracking can usually be prevented by proper conduct of the quenching operation, regardless of hardenability. Therefore, a maximum hardenability limit for this purpose is not a necessity unless controlled quenching is impossible.

Where it is impossible to modify the quench, and where high center hardness is not required, close control of hardenability is frequently effective in preventing cracking. An example is a tool joint (Fig. 10), a coupling for an oil well drill. When made of a through-hardening steel, this would crack on quenching. By selecting heats according to hardenability so that the part would harden at the surface but not at the center, cracking was eliminated. The required range of hardenability was much more restricted than could be obtained by ordering steel to the hardenability band. Selection of heats, therefore, was the only satisfactory solution.

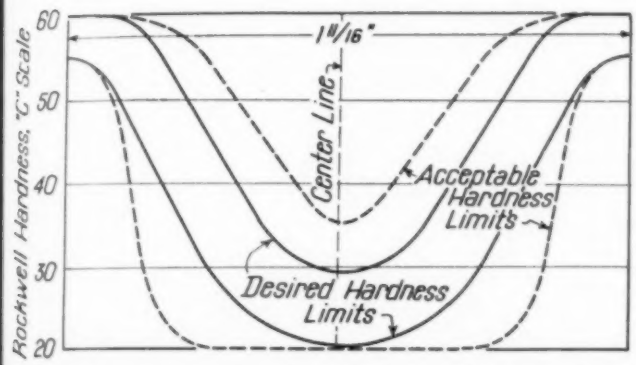


Fig. 11 — Desired Hardenability for Shell Hardened Axle Shafts, Met by Selected Heats of "10 T 40" Steel — C 1040, Grainal Treated. (Timken-Detroit Axle Co.)

Shell Hardened Axle Shafts — As the subject of residual stresses becomes better understood and efforts are made to make them work for us, it will be seen that there is more and more reason for narrow hardenability limits with a definite maximum as well as minimum. This may be illustrated by the practice used by R. W. Roush at Timken-Detroit Axle Co. for making rear axles for trucks and busses.

In a certain $1\frac{11}{16}$-in. diameter axle shaft, the hardness limits are C-55 to 60 at the surface and C-20 to 35 at the center. The steel used for a period prior to the war was S.A.E.1040 treated with an addition agent, and the hardenability limits that had to be maintained were C-45 at $\frac{3}{16}$ to $\frac{5}{16}$ in. on the Jominy end-quench bar. Since the war 1040 and 1046 are used with a "needling" treatment in the steelmaking which will produce the same hardness pattern. When quenched in a 10% caustic solution, a high surface hardness and low center hardness are obtained within the limits shown in Fig. 11.

Satisfactory shafts are obtained when hardness falls within the limits defined by the dotted lines. The *desired* limits are shown by the solid lines. Axle shafts from 85 heats made in the openhearth all came within these solid-line limits. When several sizes of axles are being made, control within these limits is possible by allocating each heat to the proper bar size, depending on hardenability determined on a cast bar prior to rolling.

Close hardenability control is necessary because the success of the axle depends on the stress pattern left in the shaft. (When center

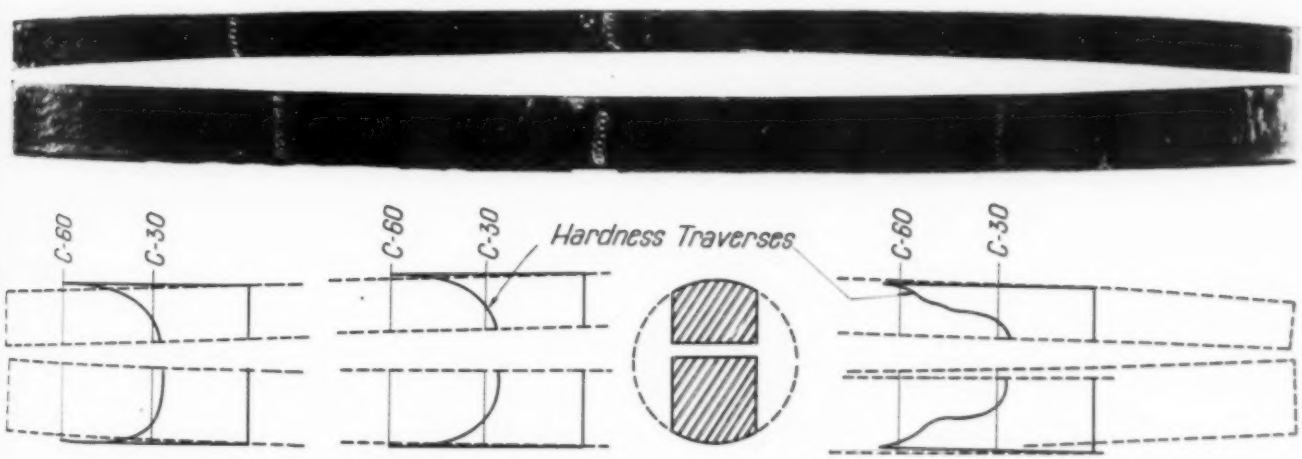


Fig. 12 — Shell Hardened Axle Shaft, Milled on Two Sides and Split Longitudinally, Bows Into a Curve From Compressive Surface Stresses. Hardness traverses at three locations. Shallow hardening steel of Fig. 11, quenched from 1550 °F. in 5% caustic solution at 110 °F., tempered at 350 °F.

hardness exceeds C-35 the shaft approaches a brittle condition so that it must be converted to a tempered shaft having a hardness of 401 to 444 Brinell.) The hard surface layers and soft center result in compressive stresses at the surface which contribute to resistance to high surface stresses from externally applied loads.

Figure 12 shows a shaft cut longitudinally; the outward bending of the two halves is evidence of the compressive stress at the surface and tensile stress at the center. Bending is sharper at one end, where it will be observed from the hardness traverses that the hardness penetration is greater. As the hardness penetration increases, the compressive stress at the surface decreases; and when full hardening occurs at the center, the stresses at the surface may be reversed and become tensile. To illustrate this, a 1-in. plate made from high hardenability steel was water quenched and then slotted down its axis. Two such pieces were placed with outside surfaces together and showed a concavity of about 0.03 in. in 6-in. span. This is proof that the surface was in tension and the cut-out center in compression.

To avoid this result is good reason for requiring a maximum limit for hardenability. On the other hand, any advantage derived from surface compressive stresses is greatest when the part is not tempered very much, because tempering reduces these stresses. Of interest is the fatigue life of such a shell hardened shaft compared to an alloy steel shaft tempered to 401 to 444 Brinell: At 60,000 psi. surface torsional stress 4340 steel lasted 10,000 cycles, 8949 lasted 36,000 and shell hardened 10T40 steel lasted 500,000 cycles!

When a shaft with high surface hardness also has high center hardness, the unfavorable trapped stresses nullify the surface strength and fatigue resistance which is a normal property of high hardness steel. A deep hardening shaft of this kind, subjected to test, breaks in a relatively few cycles with the extremely brittle type of fracture producing many small fragments near the region where the shaft finally let go. One shaft so tested (Fig. 13) had a surface hardness of C-56 and a center hardness of C-52. High-side hardenability limit, therefore, is imperative in this part, in which respect it differs from the handling of the torsion bar spring.

Carburized Parts — Rear axle ring gears, drive pinions, transmission gears, bearing races and all such highly stressed, high-hardness parts can be greatly benefited by close control of hardenability. The prime reason for this, again, is to prevent unfavorable trapped stresses, rather than to insure certain microstructures and hardness.

Success of carburized rear axle gears depends also upon a highly uniform shape of teeth to insure uniform load distribution, along with high residual compressive stresses in the case to oppose the stresses from externally applied loads. We have seen that, to obtain favorable stresses in shell hardened axles, it is necessary to control hardenability within narrow limits. The same is true for carburized gears. This same control of hardenability will also assist in uniformity of size change (distortion from hardening) which can be compensated for when cutting the gear teeth in the green. The volume changes associated with transformation in the case and core, therefore, must be held as uniform as possible, both as to magnitude and sequence.

Experience has indicated that a range of 10 points Rockwell "C" in the core should not be exceeded for good results. Core hardnesses ranging from C-25 to 35 at the root of the tooth are considered the outside desired limits. In this connection it is interesting to look at the derived cross-section hardness curves for the upper and lower hardenability limits for 4620H steel in Fig. 14. Only in sections bigger than 2-in. diameter can anything like a 10-point Rockwell range be realized. In half-inch sections, the hardness ranges from C-22 to 48.

As core hardness of gear teeth exceeds C-40, the compressive stresses desired in the case

Fig. 13 — Fragmental Type of Fatigue Fracture in Thru Hardened Axle Shaft, C-56 at Surface, C-52 at Cen

decrease, and they are replaced by tensile stresses that may build up to the cracking point of the steel. Figure 15 shows a tooth out of a gear where every tooth cracked. Its core hardness was C-40 to 45. (High core hardness is especially bad in parts where grinding must be done after hardening.) In this gear, the case cracked because the relative amounts of case and core were such that the core was stronger than the case. However, with a high ratio of case to core and a thin section part, the case is the stronger; when the core expands during hardening, it will set up tensile stresses in the case, but the case will be strong enough to assume these stresses without failing, although it may stretch some. Then when the core contracts again, after having transformed, it is subject to tri-axial stresses and cracks in a brittle manner.

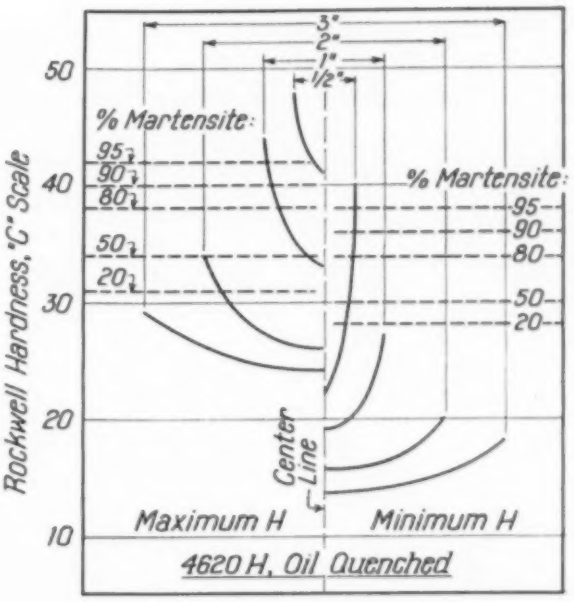


Fig. 14 — Hardness Traverses of 4620 H at Upper and Lower Limits of Hardenability Band

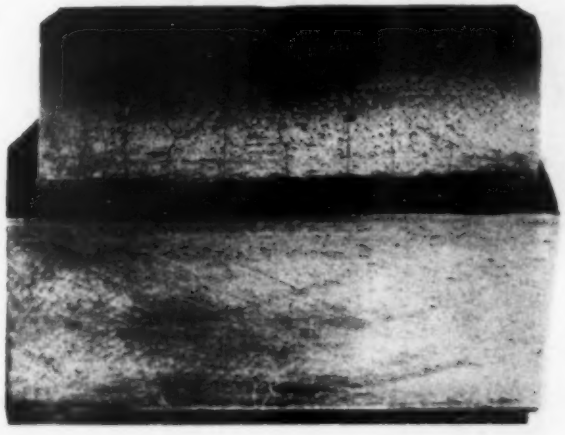


Fig. 15 — Cracked Tooth From a Gear Made of Steel With Excessive Hardness in the Core

Similar troubles are encountered in hardening roller bearing races. Radial cracks appeared when the core hardness was over C-40. Figure 16 shows a section of such a race wherein the case has cracked away from the core.

To prevent cracking such as just described — or, what is worse, high residual stresses on the verge of cracking — at least one plant specifies a hardenability band for rear axle gears in which the spread between high and low side of the hardenability band is only about half the width of the band for 4620H steel.

Fig. 16 — Sectioned Roller Bearing Race of Improper Hardness Pattern; Case Has Cracked Away From the Core

Conclusions

The analysis presented in this paper leads to conclusions as follows:

1. Ordering steel by the H-band offers opportunities for obtaining desired results at lower cost.

2. There are no instances where it is an advantage to order steel to the chemistry specification rather than the H-band specification.

3. Ordering steel to chemistry limits *plus* a minimum hardenability offers no advantage not obtained when the H specification is used, but incurs the lack of control on the high side of the band which tends toward trouble with cracking and distortion.

4. Where close hardness limits in the hardened article are necessary, the H-band limits may be too wide. Special specifications and, consequently, greater costs are involved in obtaining these special results.

5. Wherever these narrower limits are not required, the H-band specification should be used instead of a straight chemistry specification. This will focus the attention of the steelmaker on the problem of meeting hardenability limits, and will tend to afford the steel user narrower limits.

Personals

Arnold F. Meyer, for many years chief engineer of The Heil Co., Milwaukee, has been appointed vice-president in charge of engineering.

Dale H. Burke, formerly with Libbey Glass Division of Owens-Illinois Glass Co., has joined the Champion Spark Plug Co., Toledo, Ohio, as tool engineer.

Chicago Steel Service Co. announces that **Jay W. Stranahan** has been appointed district representative in the Minneapolis-St. Paul area. Mr. Stranahan was previously with Washington Steel Corp.

Peter E. Rentschler, president and treasurer of the Hamilton Foundry & Machine Co., Hamilton, Ohio, will be the recipient of the American Foundrymen's Association's Peter L. Simpson medal for his outstanding work in promoting better foundry housekeeping and safety practices, and in public relations.

Charles Locke, formerly chief metallurgist and foundry superintendent for the West Michigan Steel Casting Co., has joined the metals research department of Armour Research Foundation, Chicago.

Armour Research Foundation of Illinois Institute of Technology announces that **Ray D. Hayworth**, who was previously on the metals research staff, will now serve as supervisor of abrasion research.

H. J. Kicherer, formerly manager of the shop and assembly operations, industrial power engineering department of International Harvester Co., has been promoted to assistant chief engineer in charge of crawler tractor, industrial wheel tractor and power unit design sections.

T. M. Nolan, formerly quality manager of the Burlington, Vt., plant of Bell Aircraft Corp., has been transferred to the Buffalo division where he will be responsible for the laboratories and process engineering department as chief of laboratories.

J. R. Douglass has been promoted by Trans-Canada Air Lines from chief inspector of aircraft maintenance and overhaul to supervisor of inspection. His headquarters will be in Montreal, Quebec, Canada.

C. W. Butts is now sales representative for Fairmont Aluminum Co. for Texas and Oklahoma.

S. C. Shapleigh and **Francis J. Weldon** are operating the new Bridgeport, Conn., warehouse of the Carpenter Steel Co.

Creston E. Kite, formerly vice-president in charge of sales of General Alloys Corp. and sales executive of E. F. Houghton & Co., has joined Abarry Steel Co., Perth Amboy, N. J., as general sales manager.

Harold Smallen, formerly of Menasco Mfg. Co., Burbank, Calif., is now with the Navy at Michelson Laboratory, Inyokern, Calif.

Gerald McClure, formerly with Pittsburgh Chemical Lab., has returned to the Dubois Co. as sales representative for chemical cleaning processes for steel and rubber industries.

W. L. Ducker has closed his consulting engineering practice in Tulsa, Okla., to accept the newly established position as head of the petroleum engineering department at Texas Technological College, Lubbock, Texas.

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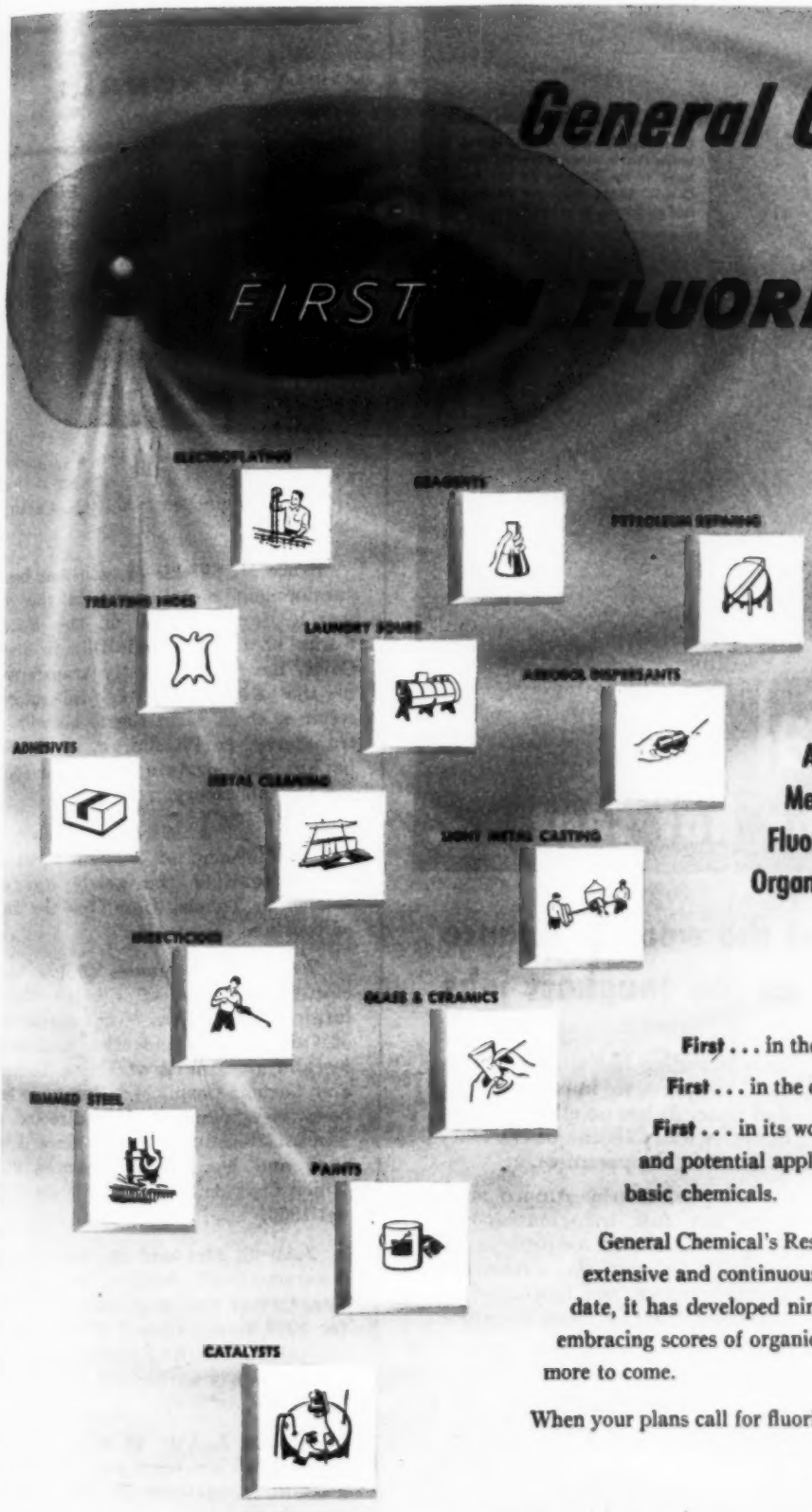

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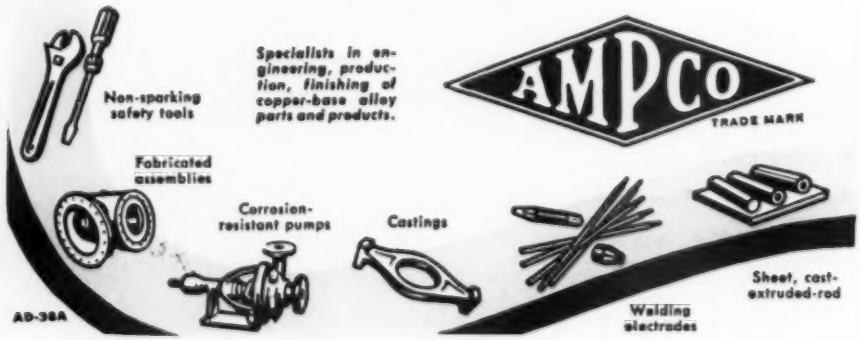

Personals

R. R. Tatnall has been appointed field service metallurgist in charge of customer relations for the Colorado Fuel and Iron Corp., Wickwire Spencer Steel Division. Mr. Tatnall has been connected with the company's Morgan plant at Worcester, Mass., for 20 years, most recently as quality control manager.

In addition to his position at Sheffield Iron & Steel Co., **S. T. Jaswinski** has recently become vice-president of the Detroit Steel Casting Co.

Bruce M. Shields, who has been development metallurgist in the research department of the South works of the Carnegie-Illinois Steel Corp., has been recently transferred to the Kearny, N. J., laboratory. After a short term there, he will be transferred to Pittsburgh, which will be his headquarters while traveling for the laboratory.

Siv Steel & Wire Co. announces the appointment of **William P. Stead** as manager of the newly opened branch in Toledo, Ohio, for the sale of tool and die steels.

Walter E. Remmers has been elected president of Electro Metallurgical Co., Electro Metallurgical Co. of Canada, Ltd., and other associated metallurgical units of Union Carbide and Carbon Corp. Mr. Remmers has been vice-president and director of Electro Metallurgical Co. since April 1945 and has been associated with Union Carbide and Carbon since August 1936.

John L. McCloud, head of the department of chemical engineering, Ford Motor Co., has been nominated for 1948 vice-president of the Society of Automotive Engineers, representing the newly authorized Engineering Materials Activity.

Marvin J. Udy announces the opening of his office for practice as consulting engineer in the field of process metallurgy, in the production of ferro-alloys, phosphorus and calcium carbide, and in general electric furnace smelting.

A. J. Williamson, until recently plant manager of the Summerill Tubing Div., is now plant manager of Columbia Steel & Shafting Co., supervising the making of both bars and tubes.

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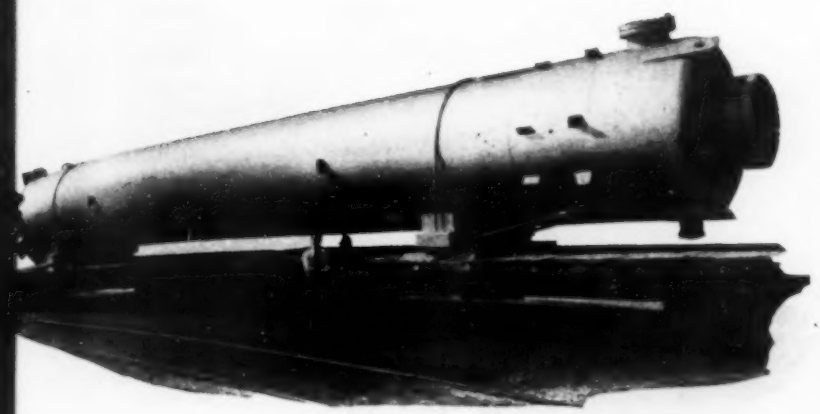
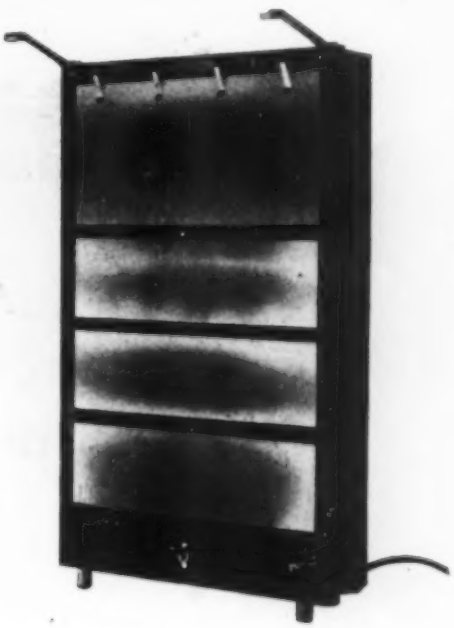


To get the facts on the internal condition of a weld . . . to provide an almost perfect control for assuring homogeneous welds . . . to give the welding engineer (and, where necessary, the customer) a permanent record of the job—use radiography.



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Radiography provides more information about the internal condition of a weld than is obtainable in any other way. It shows up irregularities, reveals lack of penetration, burned-through areas, gas and slag pockets, undercuts and other weaknesses . . . shows the welder everything he needs to know.

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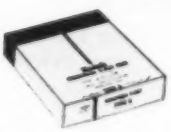

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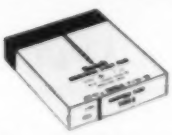

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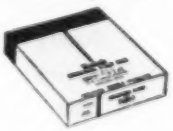

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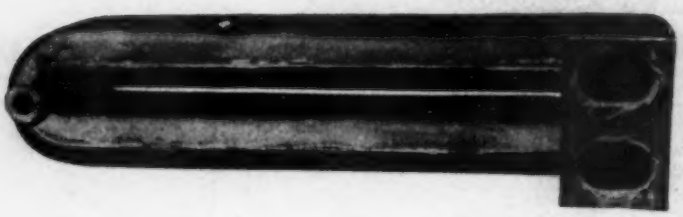

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Personals

A. R. Stevenson is now representing Mid-West Forge Co., Cleveland, in the Detroit area.

Donald H. Désy has accepted a position as metallurgist at the experimental station of E. I. du Pont de Nemours & Co., Wilmington, Del., following the completion of work toward his degree at Rensselaer Polytechnic Institute.

R. A. Wilson, formerly with the phosphate division of the research department, Monsanto Chemical Co., has recently joined the staff of the process engineering department, Air Reduction Co., Greenwich, Conn., as head of the planning and evaluation section.

R. D. Wylie has accepted a position as metallurgist at the Babcock & Wilcox Co.'s control laboratory in Barberton, Ohio. He was formerly metallurgist at the General Electric Co.'s chemical department in Pittsfield, Mass.

On graduating from the University of Pittsburgh, **Dick W. Hemphill** has become a member of the metallurgical department of the Salisbury Axle Works, Fort Wayne, Ind.

T. E. Olson, formerly metallurgist with the Chrysler Corp., is now sales engineer with the Tocco Div. of the Ohio Crankshaft Co., Cleveland.

Keller Phillippe has changed from melting department of Warman Steel Casting Co. to the sales department of the newly organized Los Angeles warehouse of the U. S. Steel Supply Co.

T. A. Read, formerly chief of the applied physics section of the Frankford Arsenal Ordnance Laboratory, has joined the physics division of the Oak Ridge National Laboratory. He continues his connection with Frankford Arsenal as a consultant.

Capt. Richard D. Fuller, who has recently returned from an overseas tour of duty, is now assigned to the Seacoast Branch of the Artillery School where he is teaching electronics in the department of engineering.

Edward S. Smith has joined the National Gypsum Co. as industrial sales representative on his graduating from the Virginia Polytechnic Institute.

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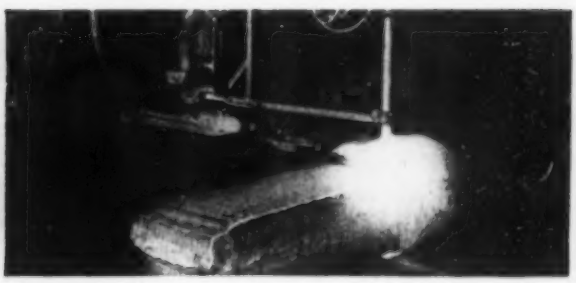

Cutting the hot top off this 26-inch forged slab with the C-45 blowpipe took less than 10 minutes and left the slab end square and undeformed.

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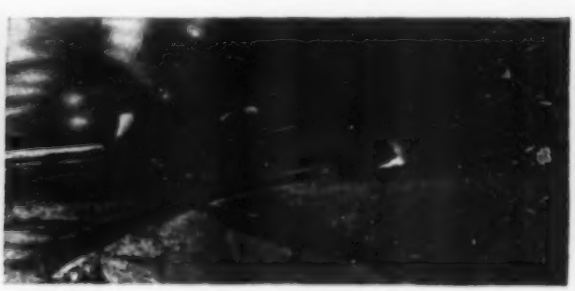

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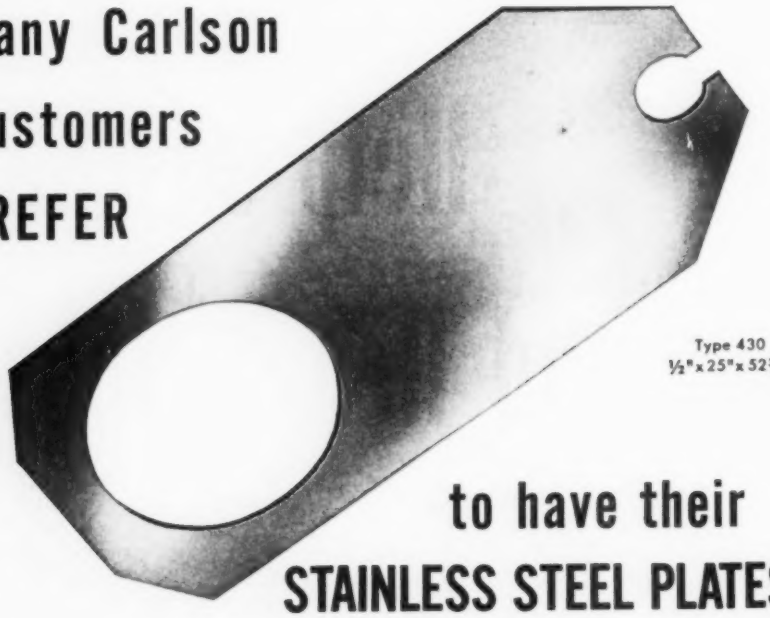

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Personals

M. O. Wirth, previously metallurgist at U. S. Naval Ordnance Plant, Center Line, Mich., has joined Ainsworth Mfg. Corp., Detroit, in the same capacity.

Ralph E. Herzler, Jr., is now a technical service engineer for Reynolds Metals Co. in Chicago.

After graduating from Rensselaer Polytechnic Institute, **Richard C. Tietze** is now employed as a methods engineer at Revere Copper & Brass, Inc., Rome, N. Y.

W. N. Crout is now district manager for the American Rolling Mill Co. in Pittsburgh.

Thomas J. Snodgrass, a graduate of Illinois Institute of Technology, has been appointed plant metallurgist of the Trenton works of the General Electric Co.

O. W. Reen, formerly metallurgist for Allegheny Ludlum Steel Corp., is now on the staff at Purdue University, Lafayette, Ind.

George L. Romine has been transferred by S. K. Wellman Co. from the Detroit office to Cleveland, where he will be district manager of original equipment sales.

Jack I. Medoff, formerly with Worthington Pump & Machine Corp. and the New York Naval Shipyard Laboratory, is conducting a general metallurgical consulting service in New York City, with laboratory facilities for testing, research and development in physical metallurgy, special attention being given to precision investment casting.

Paul K. Zimmerman, formerly special steels representative for Joseph T. Ryerson & Son, Inc., at Rock Island, Ill., has been transferred to the Alloy Steel Division at Chicago.

Earl F. Troup has been transferred as sales representative from the Buffalo to the Rochester office of Edgar T. Ward's Sons Co.

American Brass Co. has transferred **James T. Kemp**, metallurgical engineer, from Waterbury, Conn., to San Francisco.

Harold C. Stone, who started with Le Tourneau in 1937 as a heat treater, has recently been appointed metallurgical supervisor and heat treat engineer of the company.

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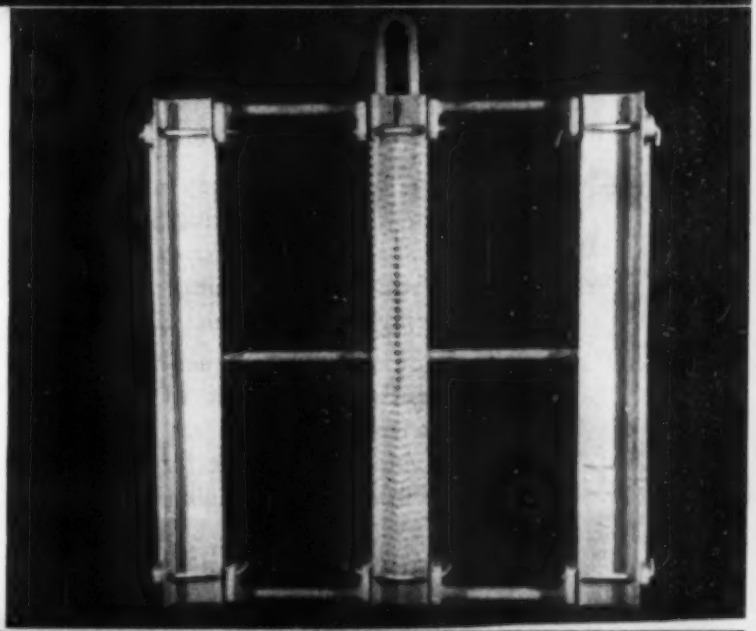

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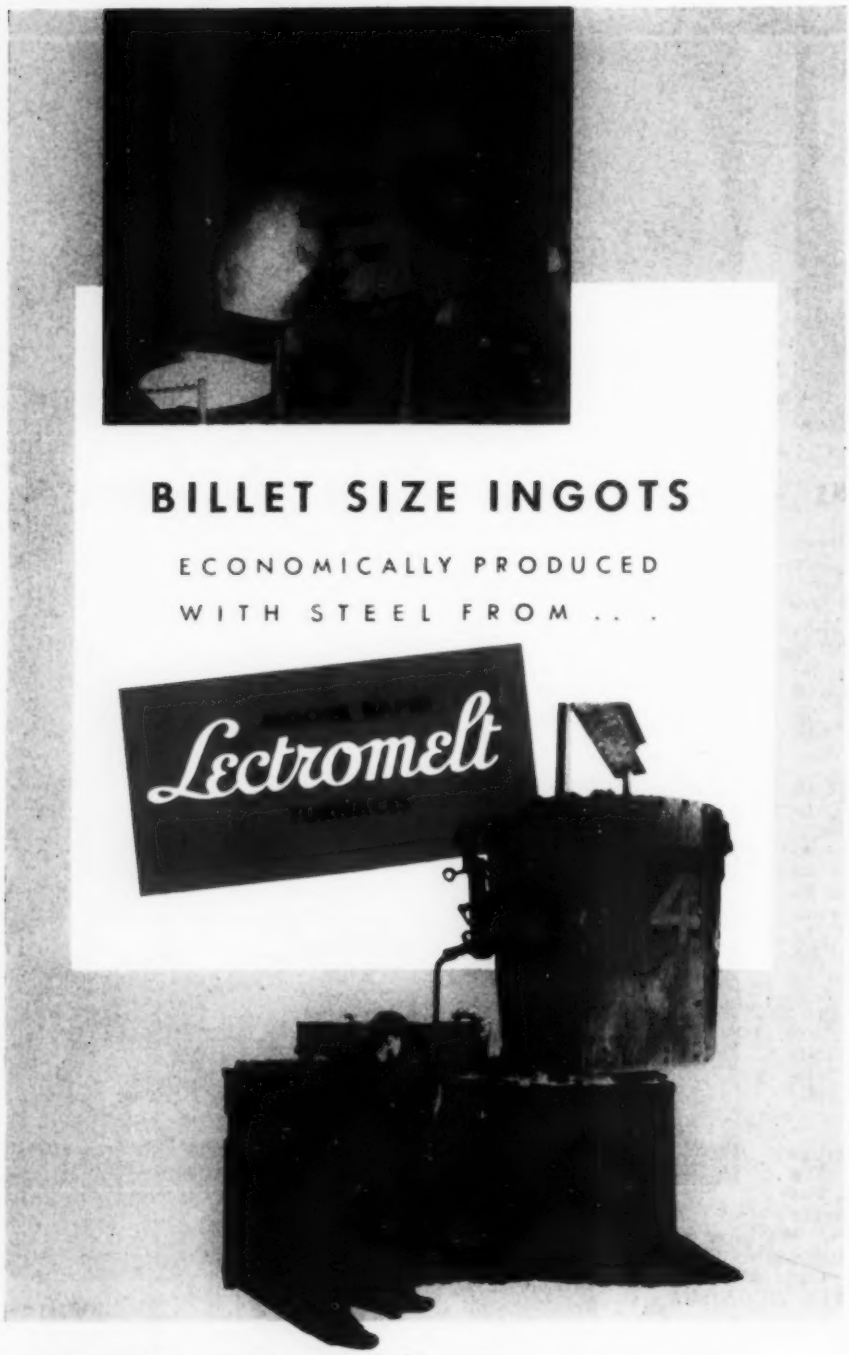

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Heat Treating in Vacuum*

EXPERIMENTS were run in two horizontal cylindrical, air-tight chambers built of 35-15 alloy tubes. Both were sealed in identical positions within a furnace body heated by three burners, the temperatures being controlled partly by checking the input of gas, partly by introducing an excess of air, so as to keep the gas flow uniform.

One of the chambers was used as a vacuum vessel, the second was operated simultaneously with a controlled atmosphere. A very high vacuum, of the order of one millionth of an atmosphere, was rapidly reached and easily maintained.

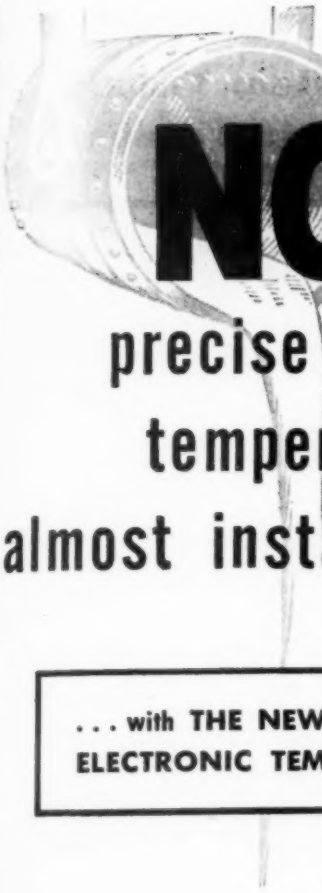
One series of experiments dealt with the problem of malleableizing. Original material contained 2.54% C, 1.26% Si, 0.47% Mn. Durations were *(a)* normal cycle, *(b)* three-quarters normal and *(c)* half-normal. Results show that large content of hydrogen in the controlled atmosphere delays malleableizing. Nitrogen has no effect and "DX" gas (5% CO_2, 10% CO, 12% H_2, 3% H_2O, 70% N_2) seems to yield higher strength and but a slightly lower elongation in malleableized test pieces.

Carburizing experiments with a mixture of nitrogen and propane carefully cleansed from all traceable oxygen have shown that the theory is fallacious which assumes oxygen to be necessary. Carburization proceeds even in somewhat better manner in its absence.

Elimination of occluded gases from deep drawing steels (meaning, essentially, hydrogen) proceeds identically in a vacuum and in nitrogen (or DX gas). The resulting Erichsen values are very close, which seems to indicate that a high degree of degassing (in vacuum) leads to no substantial improvement in drawability. Both a rimmed and a killed low-carbon steel behaved in a nearly identical manner. In both steels drawability of pickled sheet improved to the same extent after a 3-day sojourn in air at ordinary temperature.

Spheroidizing of carbides in a high-carbon steel proceeds identically in a vacuum and in a controlled atmosphere. *(To page 724)*

*Abstract of "Research in Controlled Vacuum Heat Treating" (report of work sponsored by American Gas Assoc.) by Jack Huebler.



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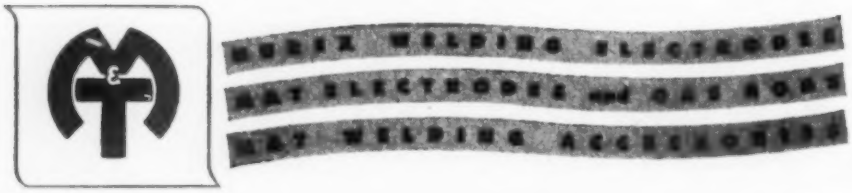

Heat Treating in Vacuum

(Continued from p. 720)

The most important types of heat treatment — hardening and drawing — proceed identically in a vacuum and in an "RX" atmosphere (20% CO, 40% H_2, 40% N_2). Neither final hardness nor impact strength change noticeably.

Heating in vacuum does not affect endurance of springs substantially.

The rates of carbon diffusion in steel remain identical in vacuum and in nitrogen. No tests with other controlled atmospheres were undertaken, but the author developed an equation which, according to him, should permit the computation of that rate under specific conditions, but no examples of completed computations are given. The equation was developed on the basis of a number of preliminary assumptions and suffers from the necessity of computing a number of factors.

Electrodeposition of Tungsten Alloys*

HARD, dense, corrosion resistant deposits of tungsten, alloyed with iron, cobalt or nickel, can be obtained electrolytically. Those with cobalt, containing 30% tungsten, deposit most easily and give the best protection to the iron below. With the same thickness of deposit, Co-W resists salt-spray better than nickel.

The alloy as precipitated is lamellar and brittle. Heating to 2375° F. eliminates this layered structure and renders the coating ductile. Initial hardness of 400 to 700 Vickers need not be lost, but can even be raised by 100 points if a drawing treatment is given at 1100° F. The deposit retains a hardness of 290 Vickers at 1300° F. — compare with 50 for a carbon steel at the same temperature. Such coating should be of service for steel parts at high temperatures.

An example of the composition of the electrolyte is given and it is added that it has very good throwing power. No further details regarding the process are included.

*Abstract of Technical Report No. 1152, National Bureau of Standards.